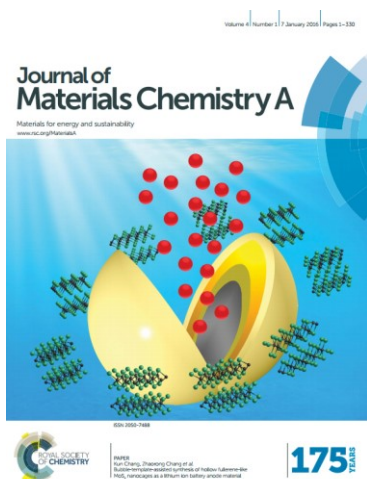


**A and B site doping of a phonon-glass perovskite oxide thermoelectric**

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Best wishes,

Simon Neil  
Managing Editor, *Journal of Materials Chemistry A*

## Authors response to reviewers comments for Journal of Materials Chemistry A article TA-ART-04-2018-003739

### Reviewer 1

No response needed

### Reviewer 2

- (1) The comment about the thermal conductivity being closely related to the microstructure is valid, however, previous work on  $\text{SrTiO}_3$  has shown that microstructural effects influence  $\kappa_{\text{latt}}$  only once nanometre-sized grains are achieved. The phonon mean free paths in  $\text{SrTiO}_3$  were shown previously (as we also show in figure 4 c)) to be very small (2-3 nm) and will be affected by grain sizes of a similar order of magnitude.

We have taken SEM images of both polished and unpolished surfaces of the pellets of  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  which are now included in the supplementary information (Fig. S6, page S8). The images show that grain sizes in these materials are of the order of 10-20 microns in size which are much larger than the phonon mean free paths that we've determined experimentally and have shown computationally through phonon calculations. As such, the effects of microstructure on  $\kappa_{\text{latt}}$  are negligible in this case. The low thermal conductivities exhibited by these compositions are intrinsic to the material due to the effects of cation disorder operating over very small length-scales, and do not result from microstructural effects.

We have added the text "Microstructure was shown to influence  $\kappa_{\text{latt}}$  and  $l_{\text{ph}}$  of  $\text{SrTiO}_3$ ,<sup>16</sup> however, this required nanometre-sized grains. The effect of microstructure in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  materials is negligible due to the large disparity between  $l_{\text{ph}}$  and grain size, which are shown through SEM (Fig. S6, ESI) to be much larger and of the order of 10-20  $\mu\text{m}$ ." to the discussion of thermal conductivity in the main text on page 7.

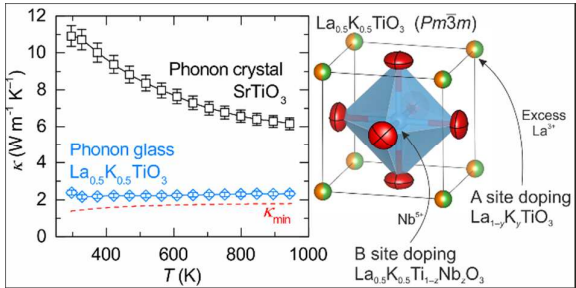
References to ESI figures in the main text have been updated on pages 8 and 9 following the reordering of some ESI figures.

- (2) Thank you for pointing this out. This has now been corrected.

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**A and B site doping of a phonon-glass perovskite oxide thermoelectric**

The effect of structural symmetry is investigated in phonon-glass electron-crystal (PGEC)  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  thermoelectric oxides





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## ARTICLE

# A and B site doping of a phonon-glass perovskite oxide thermoelectric

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By tuning the A site cation size it is possible to control the degree of octahedral distortion and ultimately structural symmetry in the new perovskite solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ , affording a rhombohedral-to-cubic transition as  $x$  increases above 0.4. The  $\text{La}^{3+}$  and  $\text{K}^+$  cations are distributed randomly across the A site leading to significant phonon disorder in cubic  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  ( $Pm\bar{3}m$ ) which produces a phonon-glass with a thermal conductivity of  $2.37(12) \text{ W m}^{-1} \text{ K}^{-1}$  at 300 K; a reduction of 75% when compared with isostructural  $\text{SrTiO}_3$ . This simple cation substitution of  $\text{Sr}^{2+}$  for  $\text{La}^{3+}$  and  $\text{K}^+$  maintains the flexible structural chemistry of the perovskite structure and two mechanisms of doping for the introduction of electronic charge carriers are explored; A site doping in  $\text{La}_{1-x}\text{K}_x\text{TiO}_3$  or B site doping in  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$ . The phonon-glass thermal conductivity of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is retained upon doping through both of these mechanisms highlighting how the usually strongly coupled thermal and electronic transport can be minimised by mass disorder in perovskites. Precise control over octahedral distortion in A site doped  $\text{La}_{1-x}\text{K}_x\text{TiO}_3$  which has rhombohedral ( $R\bar{3}c$ ) symmetry affords lower band dispersions and increased carrier effective masses over those achieved in B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  which maintains the cubic ( $Pm\bar{3}m$ ) symmetry of the undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  parent. The higher Seebeck coefficients of A site doped  $\text{La}_{1-x}\text{K}_x\text{TiO}_3$  yield larger power factors and lead to increased thermoelectric figures of merit and improved conversion efficiencies compared with the mechanism for B site doping.

## Introduction

Thermoelectric devices convert heat energy directly into electricity or vice versa and have been utilised heavily as power sources for spacecraft and in thermoelectric heating or cooling technologies.<sup>1, 2</sup> The most important application of these devices is to the recovery of waste heat energy, particularly in the industrial, residential and automotive sectors in order to improve efficiency and sustainability.<sup>3, 4</sup> The performance of thermoelectric materials is assessed via the thermoelectric figure of merit  $ZT = (S^2\sigma/\kappa)T$  comprised of the electronic conductivity ( $\sigma$ ) and Seebeck coefficient ( $S$ ), which together form the power factor ( $S^2\sigma$ ), and the thermal conductivity ( $\kappa$ ). Optimisation of  $ZT$  is made difficult by the complex coupled relationships that exist between each of the above properties. The phonon-glass electron-crystal (PGEC) approach envisages the decoupling of  $\sigma$  and  $S$ , which are

typically governed by the Boltzmann transport equation, from  $\kappa$  which is governed by phonon transport within the lattice.<sup>5</sup> "Phonon-glass" refers to the lattice contribution to the thermal conductivity ( $\kappa_{\text{latt}}$ ) being minimised such that heat transport in a material is hindered like in a glass, while "electron-crystal" corresponds to the retention of a high  $\sigma$  and  $S$  within the same material leading to charge transport like in a semiconductor. These characteristics are widely reported in intermetallic systems such as clathrates and skutterudites,<sup>6, 7</sup> which rely on the inclusion of rattler guest atoms within large structural cages to suppress  $\kappa_{\text{latt}}$  while electronic transport mediated through the covalent framework remains unaffected. The PGEC concept has been recently extended to oxide thermoelectrics.<sup>8</sup>

Metal oxides offer increased stability at high operating temperatures (automotive exhaust waste heat recovery typical operation temperature is 350-700 K, and industrial furnace waste heat recovery at 700-1100 K)<sup>4, 9</sup> as well as higher natural abundance and lower toxicity of the constituent elements, leading to much lower costs and smaller environmental impacts when compared with the intermetallics used currently.<sup>10, 11</sup> Titanate-based thermoelectrics, such as donor doped  $\text{SrTiO}_3$ , are the current state-of-the-art n-type thermoelectric oxides with a recent report of  $ZT \approx 0.6$  at 1000 K which was achieved through nanoscale modulation of  $\text{SrTiO}_3$

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<sup>†</sup> Electronic Supplementary Information (ESI) available. See DOI: 10.1039/x0xx00000x

and  $\text{TiO}_2$  inclusions in  $\text{Sr}_{0.9}\text{La}_{0.1}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$ .<sup>12</sup> The unique band structure of  $\text{SrTiO}_3$  with contributions from both heavy and light carriers at the Fermi level,<sup>13</sup> gives rise to high power factors of  $28\text{--}36\text{ mW K}^{-2}\text{ cm}^{-1}$  at 300 K for  $x = 0.015\text{--}0.1$  in  $\text{Sr}_{1-x}\text{La}_x\text{TiO}_{3-\delta}$  single crystals but  $ZT$  is limited to 0.09 due to the large  $\kappa$  of  $9\text{--}12\text{ W m}^{-1}\text{ K}^{-1}$  in these materials.<sup>14, 15</sup> With  $\kappa_{\text{latt}}$  contributing 75–100% of the total thermal conductivity, control over lattice phonon transport would allow for much higher figures of merit to be achieved in titanate perovskite thermoelectrics.  $\text{SrTiO}_3$  can be considered as a phonon crystal where heat transport is mediated by coherent and long-distance propagation of phonons through the lattice, but due to the relatively small phonon mean free path (2–3 nm) the effects of nanostructuring approaches that are widely used to reduce  $\kappa$  in intermetallic thermoelectrics are negated unless nanometre-sized grains are achieved.<sup>16</sup> We highlighted previously the effectiveness of targeting a low  $\kappa$  that is intrinsic to the material, as opposed to originating from extrinsic microstructural effects, by exploiting the flexibility of the perovskite structure in a simple A site cation substitution that lead to a phonon-glass  $\kappa$  of  $\approx 2.2\text{ W m}^{-1}\text{ K}^{-1}$  in crystalline  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ .<sup>8</sup> Through a combined experimental and computational approach, the origin of this glass-like behaviour was explained by increased phonon scattering as a result of charge and mass disorder resulting from the random distribution of  $\text{La}^{3+}$  and  $\text{Na}^+$ . Specifically, compared against  $\text{SrTiO}_3$  the disordered cations in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  act as locally decoupled oscillators with a broad range of vibrational frequencies which ultimately lead to phonon interactions being far more localised and heat transport being much less coherent. By achieving this intrinsically low  $\kappa$ , it enabled the emergence of PGEC properties where thermal and electronic transport are decoupled and could be altered independently.

The unique band dispersions observed in  $\text{SrTiO}_3$  that allow for high power factors are correlated significantly with the high symmetry of the cubic structure and linearity of the Ti–O–Ti bonding.<sup>13</sup>  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  is a rhombohedrally-distorted perovskite with an  $\approx 8^\circ$  tilt of the  $\text{TiO}_6$  octahedra and non-linear Ti–O–Ti bond angles of  $\approx 167^\circ$ .<sup>8, 17</sup> This tilting decreases overlap interactions between the Ti 3d and O 2p orbitals that are responsible for the dispersion of the valence and conduction bands in  $\text{SrTiO}_3$ , which decreases band dispersion thus increasing the charge carrier effective mass and in turn reducing the electronic conductivity. Symmetry considerations indicate that band dispersions are highest in the cubic perovskite phase, leading to the highest conductivity.<sup>8, 13</sup> It is a widely accepted feature of the perovskite structure that it is possible to influence the symmetry of the material by tuning the cation size, and the effect of this can be estimated through the Goldschmidt tolerance factor,  $t = (r_A + r_O)/\sqrt{(r_B + r_O)}$ , where the highest symmetry cubic structures have values closest to unity.<sup>18, 19</sup> By increasing the cation size on the A site of  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  through substitution of  $\text{Na}^+$  for  $\text{K}^+$  ( $r_{\text{Na}^+} = 1.39\text{ \AA}$  and  $r_{\text{K}^+} = 1.64\text{ \AA}$ )<sup>20</sup> in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ ,  $t$  can be increased from 0.98 closer to 1. This is a new solid solution that is not reported in the literature and very little information is available on the fully substituted end-member phase

$\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . This phase has been prepared previously through hydrothermal,<sup>21</sup> and high-pressure synthesis,<sup>22</sup> however detailed analysis and understanding of the structures is lacking. A cubic structure was reported by Brous *et al.*,<sup>23</sup> however this was based on a non-stoichiometric composition of  $\text{La}_{0.531}\text{K}_{0.469}\text{TiO}_3$ . We report the ceramic ambient pressure synthesis and detailed study of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  and the new perovskite solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . Through crystal structure engineering and tuning of A site cation size it is possible to obtain cubic symmetry leading to band dispersions optimised for maximum conductivity in the perovskite  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  while retaining the A site disorder that suppresses  $\kappa$  and leads to phonon-glass behaviour. The chemical flexibility of this material is demonstrated by donor doping on either the A or B site giving rise to electronic transport and power factors comparable to other doped titanate perovskites while the thermal conductivity remains mostly unaffected.

## Experimental

### Synthesis and Processing of $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$

Pre-dried reagents of  $\text{La}_2\text{O}_3$  (99.99%, Aldrich),  $\text{Na}_2\text{CO}_3$  (99.997%, Alfa Aesar),  $\text{K}_2\text{CO}_3$  (99.997%, Alfa Aesar) and  $\text{TiO}_2$  (99.99%, Aldrich) were weighed out on a  $\approx 1.7$  gram scale to give desired compositions. All reagents were dried overnight at 473 K before weighing, except for  $\text{La}_2\text{O}_3$  which was annealed at 1223 K. The powder mixtures were ground in acetone for 10 minutes before being fired as a powder in air at 1273 K for 4 hours placed in an alumina crucible. Heating and cooling rates were 1 and 5  $\text{K min}^{-1}$ , respectively. The fired powders were ground for 10 minutes before being pressed uniaxially into a 13 mm pellet approximately 1 gram in mass. A portion of the powder was kept separate ( $\approx 0.7$  grams) to be used as sacrificial mixture in order to minimise the loss of volatile  $\text{Na}_2\text{O}$  and  $\text{K}_2\text{O}$  species during sintering. The pellet was then further pressed under an isostatic pressure of 200 MPa at ambient temperature using an Autoclave Engineers Cold Isostatic Press. These cold-pressed pellets were placed in an alumina boat and buried in the sacrificial powder before being sintered for 8 hours at 1623 K in air using heating and cooling rates of 1 and 3  $\text{K min}^{-1}$ , respectively.

### Synthesis and Processing of $\text{La}_{1-y}\text{K}_y\text{TiO}_3$ and $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$

Preparation of powder mixtures for the doped materials was done using a procedure similar to that described above for  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ .  $\text{Nb}_2\text{O}_5$  (99.9985%, Alfa Aesar) was used for the B site doped compositions. The sintering step performed at 1623 K for 8 hours, with the cold-pressed pellet buried under sacrificial powder of the same composition, was done under a  $50\text{ mL min}^{-1}$  flow of  $\text{H}_2/\text{N}_2$  (5/95%) instead of air in order to reduce the sample. To ensure the processing conditions were consistent for each composition, the  $p(\text{O}_2)$  was monitored during each reduction step using a Cambridge Sensotec Rapidox 2100 Oxygen Analyser. The furnace was purged with flowing  $\text{H}_2/\text{N}_2$  (5/95%) to achieve a  $p(\text{O}_2)$  of  $10^{-19}$  ppm before the pellets were sintered.

Pairs of pellets for each composition were processed in the same reactions; one pellet for the measurement of thermal transport, and the other for measurement of electronic transport. Pellet densities were determined using the Archimedes method.

### Characterisation

Structural analyses of the new  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ ,  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  materials were performed through Rietveld analysis of synchrotron X-ray powder diffraction (SXRD) data and powder neutron diffraction (PND) data. SXRD data were recorded at beamline I11 ( $\lambda = 0.82608 \text{ \AA}$ ) at Diamond Light Source, U.K, with the powders contained within 0.3 mm diameter borosilicate capillaries. PND data were collected on HRPD at ISIS, the U.K. time-of-flight spallation neutron source, with the powders loaded into thin-walled cylindrical vanadium cans of 8 mm diameter. Data were measured at ambient conditions on detector banks 1 ( $2\theta = 168^\circ$ ), 2 ( $2\theta = 90^\circ$ ) and 3 ( $2\theta = 30^\circ$ ) and were corrected for absorption effects before analysis. Rietveld refinements were carried out using Topas Academic.<sup>24</sup> SEM images were collected using a Hitachi S-4800, cold cathode, field-emission scanning electron microscope (SEM) from pellets that were deposited on a carbon tape attached to an aluminium stub.

### Elemental analysis

Analysis of sample compositions was undertaken using Inductively Coupled Plasma Optical Emission Spectrometry (ICP-OES). Solutions of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  ( $y = 0.5, 0.45$  and  $0.4$ ) and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  ( $z = 0.05$  and  $0.1$ ) were prepared through acid digestion of approximately 10 mg pieces of sintered pellet. These were dissolved in 7 mL of concentrated HCl sealed inside Teflon-lined stainless-steel Parr autoclaves which were heated to 503 K for 4 hours. These solutions were made up to 100 mL in volumetric glassware using ultra-pure water. Metal contents were corrected for interference by comparing separate and combined metal solutions of known concentrations. Data were collected on an Agilent 5110 SCDV ICP-OES instrument.

### Measurement of thermal conductivity

Thermal diffusivity ( $\alpha$ ) data were measured by laser flash analysis using a Netzsch LFA 457. The polished pellets (approximately 10 mm in diameter and 1.5–2 mm thick) were coated in colloidal graphitic carbon before being placed inside the sample chamber which was evacuated and purged three times before being put under a constant helium flow of 100 mL min<sup>-1</sup>. Data were collected in 50 K steps in the temperature range 298 to 943 K, with a five minute equilibration at each temperature. Three measurements were recorded at each temperature and averaged and the standard deviation of these points was typically below 0.5%. Thermal expansion profiles were measured on a Netzsch DIL 402C push-rod dilatometer under dynamic vacuum on bars approximately 4 mm in length. Continuous measurements were made from 298 to 973 K with a heating rate of 3 K min<sup>-1</sup>. Errors on dilatometry measurements are assumed to be 5% as informed by the manufacturer. Heat flux profiles were measured on small pieces of sintered pellet ( $\approx 50 \text{ mg}$ ) using a Netzsch DSC 404 F1

differential scanning calorimeter under a 50 mL min<sup>-1</sup> flow of helium. Data were recorded from 323 to 973 K at a heating rate of 10 K min<sup>-1</sup>. Data were also measured under identical conditions from a sapphire standard (also 50 mg) which was used to determine the heat capacities ( $C_p$ ) of each sample. Errors on heat capacities were assumed to be 5% as advised by the manufacturer. The thermal conductivity ( $\kappa$ ) was calculated by combining the diffusivity, dilatometry and heat capacity data through  $\kappa(T) = \alpha(T) \cdot C_p(T) \cdot \rho(T)$  where  $\rho$  is the pellet density. The total thermal conductivity ( $\kappa$ ) is equal to the sum of the lattice ( $\kappa_{\text{latt}}$ ) and electronic ( $\kappa_{\text{elec}}$ ) contributions through  $\kappa = \kappa_{\text{latt}} + \kappa_{\text{elec}}$ . The electronic contribution to the thermal conductivity is estimated through the Wiedemann-Franz law which states  $\kappa_{\text{elec}} = LT\sigma$ , where  $L$  is the Sommerfeld value for the Lorenz number ( $2.45 \times 10^{-8} \text{ W } \Omega \text{ K}^{-2}$ ),  $T$  is the temperature and  $\sigma$  is the electronic conductivity.<sup>25</sup>

### Measurement of electronic conductivity and Seebeck coefficient

Simultaneous electronic conductivity ( $\sigma$ ) and Seebeck coefficient ( $S$ ) measurements were measured on an Ulvac-Riko ZEM-3 instrument from rectangular bars (dimensions of  $2 \times 2 \times 8 \text{ mm}$ ) which were cut from sintered pellets using a low-speed diamond-blade saw. The bars were mounted in a 4-point geometry with outer current electrodes in contact with each end face and inner thermocouple and voltage probes pressed against one of the longer (8 mm) sides of the bar. The sample chamber was evacuated and purged three times with helium and then dosed with 0.01 MPa of helium before commencing measurement. Data were recorded at 50 K intervals from 323 to 973 K, applying 10, 20 and 30 K temperature gradients to the bar at each temperature step. Errors on both  $\sigma$  and  $S$  were assumed to be 5% on advice of the manufacturer.

### Phonon calculations

To achieve a better understanding of the atomistic origin of the different thermal conductivities of  $\text{La}_{0.5}\text{K}_{0.5}\text{TO}_3$  and  $\text{SrTiO}_3$ , accurate first principles calculations were performed. All phonon calculations of  $\text{La}_{0.5}\text{K}_{0.5}\text{TO}_3$  (with random La/K distribution) and  $\text{SrTiO}_3$  were performed in the harmonic approximation using the code CP2K, at the density functional theory (DFT)/PBE level of theory.<sup>26, 27</sup> CP2K employs a mixed Gaussian/plane-wave basis set; in the current study, a double- $\zeta$  plus polarization quality Gaussian basis set,<sup>28</sup> and a 400 Ry plane-wave cutoff for the auxiliary grid were employed in conjunction with Goedecker–Teter–Hutter pseudopotentials.<sup>29, 30</sup> A convergence threshold of  $1.0 \times 10^{-8} \text{ Ha}$  was used for the self-consistent field cycle, and structural optimizations were considered to have converged when the maximum force on all atoms fell below  $1.0 \times 10^{-4} \text{ Ha Bohr}^{-1}$ . All calculations were performed with the  $\Gamma$ -point approximation, using a rhombohedral supercell containing 640 atoms (lattice constants  $\sim 22 \text{ \AA}$ ). The second derivatives of the total energy with respect to the atomic displacements were computed numerically, from analytical first derivatives, using atomic displacements of 0.01 Bohr from the equilibrium position. Normal modes and frequencies were obtained by diagonalization of the dynamical matrix. To understand the nature of phonon modes in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ , the mean distance

between the 10 atoms that participate to the greatest extent in a given phonon mode as defined by their displacement amplitudes were calculated, denoted as  $d_{\text{average}}$ , which quantifies the localisation of phonon modes and is given by:

$$d_{\text{average}} = \frac{1}{n} \sum_{i,j} d_{ij} \quad (1)$$

where  $i$  and  $j$  are indices of the atoms with the largest displacement amplitudes for the normal mode considered),  $d_{ij}$  is their interatomic distance (using minimum image convention), and  $n$  is the number of pairs of atoms.

#### Electronic band structure calculations

To understand how the octahedral tilting (as present in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  for  $x < 0.4$ ) modifies the band dispersions at the bottom of the conduction band, which may directly affect the thermoelectric power factor, we have performed additional DFT calculations on the electronic band structures of  $\text{SrTiO}_3$ , taking it as a model system, and we consider different amounts of octahedral tilting. All electronic band structure calculations of  $\text{SrTiO}_3$  were performed using CRYSTAL14 at DFT/PBE0 level of theory.<sup>31, 32</sup> Reciprocal space sampling has been performed for all structures using a Monkhorst-Pack grid of  $8 \times 8 \times 8$  k-points. Standard all electron basis sets from the CRYSTAL online database ([www.crystal.unito.it](http://www.crystal.unito.it)) have been used for O and Ti (indicated by the following labels online: O\_8-411d11G\_valenzano\_2006, Ti\_86-411(d31)G\_darco\_unpub). For Sr, we have used a small-core relativistic effective core potential with a double-zeta quality valence basis set (indicated by Sr\_m-S-RSC\_Heyd\_2005\_Heyd\_2005 online). The initial structure of cubic  $\text{SrTiO}_3$  in the rhombohedral cell (space group  $R\bar{3}c$ ) without any octahedral tilting has been fully optimised, which was then used as the starting point to generate  $\text{SrTiO}_3$  structures with octahedral tilting. The two  $\text{SrTiO}_3$  structures with different amounts of octahedral tilting were manually generated, by adjusting the oxygen fractional coordinates and lattice parameters, to ensure Ti–O bond lengths remain the same. Subsequent band structure calculations were performed for the unoptimized structures (for  $\text{SrTiO}_3$  with octahedral tilting).

## Results and Discussion

### Phonon-glass thermal conductivity in $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$

$\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  has a rhombohedral ( $R\bar{3}c$ ) perovskite structure, described by a tilting of the  $\text{TiO}_6$  octahedra about the [111] direction of the primitive cubic cell ( $a^-a^-a^-$  tilt system in Glazer's notation),<sup>33</sup> and disordered A site cation distribution. Phonon calculations that previously highlighted the localisation of vibrational modes in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  were performed for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Calculation of  $d_{\text{average}}$ , which corresponds to the average separation between the ten atoms with largest phonon displacement amplitudes for each individual phonon mode is shown in Fig. 1 for  $\text{SrTiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The results show that  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  has similarly promising features to those observed in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ ,<sup>8</sup>

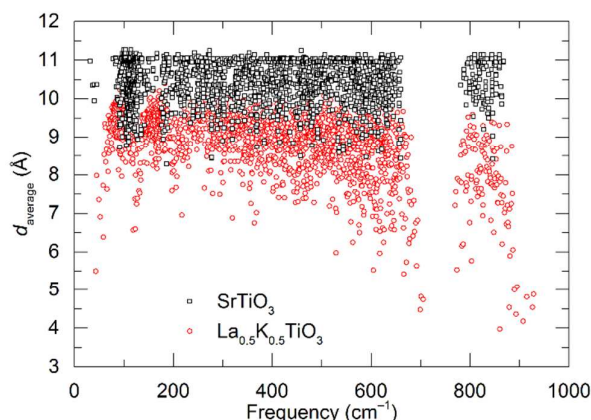


Fig. 1 Phonon localisation in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  (red circles) determined from DFT calculations compared against interactions in the phonon-crystal  $\text{SrTiO}_3$  (black squares). The parameter  $d_{\text{average}}$  is calculated for individual phonon modes as the mean distance between the ten atoms that participate to the greatest extent in that phonon mode as defined by their displacement amplitudes.

indicating that the disordered distribution of  $\text{La}^{3+}$  and  $\text{K}^+$  leads to significant phonon localisation and should yield a phonon-glass thermal conductivity similar to  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ . Compared against  $\text{SrTiO}_3$  which has a  $d_{\text{average}}$  of  $\approx 11$  Å across the entire phonon spectrum,  $d_{\text{average}}$  is reduced by 15–20% in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . We also showed previously how the coexisting dispersive and dispersion-less bands at the bottom of the conduction band that allow for high power factors in cubic  $\text{SrTiO}_3$  are retained in rhombohedral  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ . Table 1 shows relative effective masses ( $m^*$ ) extracted from the calculated electronic structures of  $\text{SrTiO}_3$  at different octahedral tilt angles. The effective masses decrease with increasing tilt angle for both bands, which shows that a cubic zero-tilt structure provides band dispersions that are optimised for highly mobile carriers towards the best possible electronic conductivity. From these computational indications,  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  presents a suitable low thermal conductivity target with band dispersions optimised for electronic conductivity and the structural flexibility of a perovskite to allow for carrier doping.

Potassium was substituted for sodium through the new solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . The results of Rietveld refinement against SXRD and PND data measured on  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ , prepared at 1300 °C through solid state synthesis, are shown in Fig. 2. The single phase powder diffraction pattern for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  can be indexed to a cubic unit cell with space group symmetry  $Pm\bar{3}m$  and lattice parameter of 3.908212(5) Å. This is slightly larger than the only reported lattice parameter of 3.907 Å,<sup>23</sup> which was based on a composition of  $\text{La}_{0.531}\text{K}_{0.469}\text{TiO}_3$  implying that more of the larger  $\text{K}^+$  is present in the structure reported here as evidenced by the refined occupancies (Table S1, ESI) and presents the first structural analysis of stoichiometric  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The  $\text{La}^{3+}$  and  $\text{K}^+$  are modelled on the  $1a$  crystallographic position and both A and B sites are modelled using isotropic displacement

Tilt angle (°)	$m^*_{\text{flat}}$	$m^*_{\text{dispersive}}$	$m^*_{\text{flat}}/m^*_{\text{dispersive}}$
0.00	6.14	1.00/1.00	6.14/6.14
3.25	8.07	1.02/0.99	7.89/8.13
6.50	14.05	1.14/1.05	12.37/13.35

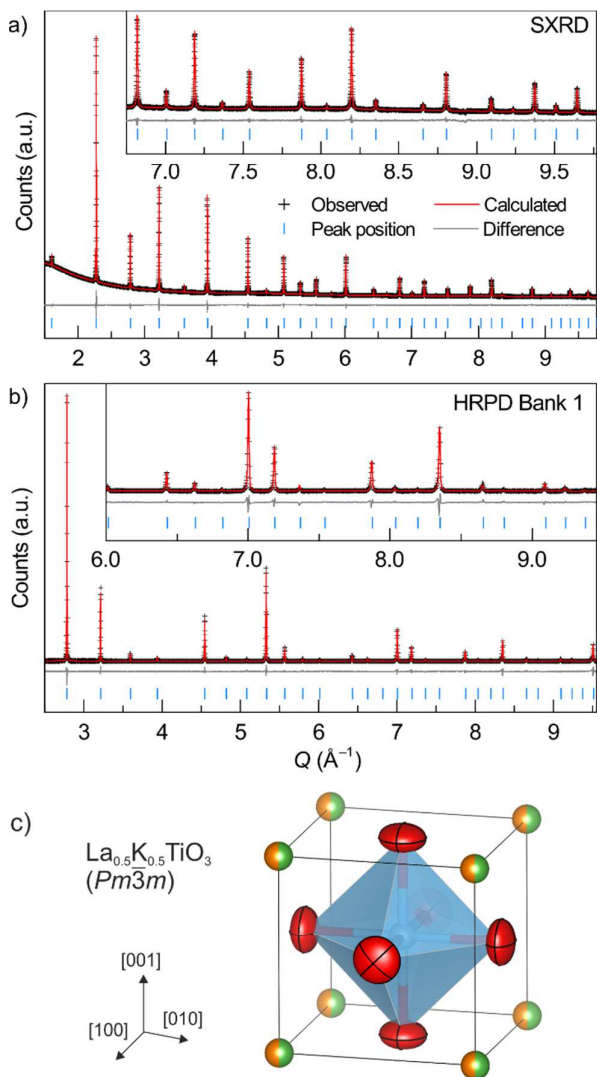


Fig. 2 Rietveld refinements against a) SXRD ( $\lambda = 0.82608 \text{ \AA}$ ) and b) high resolution NPD data measured on Bank 1 (backscattering  $2\theta = 168^\circ$ ) of HRPD at room temperature for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The refinement model is based on cubic  $Pm\bar{3}m$  symmetry with A site metals  $\text{La}^{3+}$  and  $\text{K}^+$  modelled on the  $1a$  (0, 0, 0) position, the  $\text{Ti}^{4+}$  on the  $1b$  ( $\frac{1}{2}, \frac{1}{2}, \frac{1}{2}$ ) position and oxide ions on the  $3c$  positions ( $0, \frac{1}{2}, \frac{1}{2}$ ). The cubic cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is shown in c) with anisotropic displacement ellipsoids for the oxide positions.

parameters (Table S1, Fig. S1, ESI). The use of anisotropic displacement parameters (ADPs) to model the  $3c$  oxygen position yielded a small improvement to

Table 1 Effective masses ( $m^*$ ) for the flat and two dispersive bands along Gamma-Kappa direction extracted from hybrid DFT calculations at different tilt angles in  $\text{SrTiO}_3$ . The effective masses are normalised to the effective mass of the dispersive bands in the cubic structure without octahedral tilting.

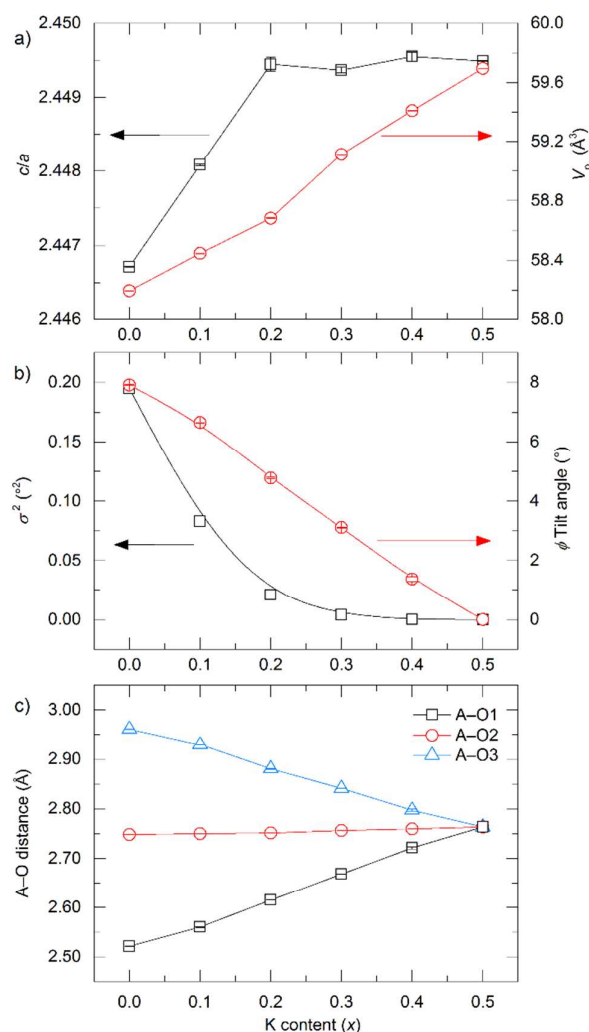


Fig. 3 Variation in a)  $c/a$  lattice parameter ratio (black squares) and primitive pseudocubic cell volume (red circles), b) bond angle variance (black squares) and octahedral tilt angle (red circles), and c) A–O distances as a function of  $x$  in the solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  refined from SXRD and high resolution NPD data. Primitive cell parameters are obtained through  $a_p = a/\sqrt{2}$  and  $c_p = c/2\sqrt{3}$ . Octahedral tilt angles ( $\phi$ ) are determined through  $\phi = \arctan[2\sqrt{3}(x-0.5)]$ , where  $x$  is the  $x$  coordinate of the  $18e$  oxide position in space group  $R\bar{3}c$ . Bond angle variance is obtained from eqn (2) in the text.

the fit, decreasing  $R_{wp}$  from 1.525% to 1.485%. This gave a similar equivalent  $U_{iso}$  ( $0.01446(16) \text{ \AA}^2$ ) to modelling the site using isotropic displacement parameters ( $0.01425(7) \text{ \AA}^2$ ) but resulted in slightly elongated ADPs displaying atomic motion perpendicular to the Ti–O–Ti direction as is expected due to thermally induced rotational motion of the  $\text{BO}_6$  octahedra.

A rhombohedral-to-cubic transition exists within the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  series and occurs upon increasing  $x$  between 0.4 and 0.5 with  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  being cubic ( $Pm\bar{3}m$ ), and compositions of  $0.1 < x < 0.4$  being indexed to the same rhombohedral unit cell as  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  ( $R\bar{3}c$ ). This rhombohedral-to-cubic transition that occurs between  $x = 0.4$  and 0.5 is not immediately obvious upon examination of the  $c/a$  lattice parameter ratio shown in Fig. 3 a) which suggests that the unit cell becomes metrically cubic at  $x = 0.2$  (where  $c/a = \sqrt{6} = 2.4495$ ), and not at  $x = 0.5$ . Only full structural analysis through Rietveld refinement can confirm that the



rhombohedral structure, defined by the symmetry elements present, persists in the  $0.1 < x < 0.4$  compositions and that cubic symmetry is achieved only for  $x = 0.5$ . The primitive pseudocubic cell volume increases linearly towards  $x = 0.5$  as shown in Fig. 3 a). The replacement of smaller sodium with larger potassium ( $r_{\text{Na}^+} = 1.39 \text{ \AA}$  and  $r_{\text{K}^+} = 1.64 \text{ \AA}$ )<sup>20</sup> increases the average cation size on the A site and causes an expansion of the unit cell. This expansion of the unit cell with increasing  $x$  results in a linear decrease in the octahedral tilt angle,  $\phi$ , which is used to describe the rotation of the  $\text{BO}_6$  octahedra about the threefold axis of the primitive cubic cell and is determined from the convention of Megaw,  $\phi = \arctan[2\sqrt{3}(x - 0.5)]$ , where  $x$  is the positional parameter of the 18e oxygen site.<sup>34</sup> Fig. 3 b) shows how the octahedral tilt angle decreases from  $7.9142(2)^\circ$  for  $x = 0$  to  $0^\circ$  for  $x = 0.5$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ , with the rhombohedral-to-cubic transition between tilted and non-tilted structures occurring between  $x = 0.4$  and  $0.5$ . Alternatively, this can be viewed through the Ti–O–Ti bond angles which become more linear with  $x$ , increasing from  $167.044(14)^\circ$  for  $x = 0$  to  $180^\circ$  in  $x = 0.5$ . While  $\phi$  is decreasing towards  $0^\circ$  with increasing  $x$ , there is a simultaneous distortion of the octahedra themselves where O–Ti–O angles deviate away from the ideal  $90^\circ$  and can be quantified by the octahedral bond angle variance in eqn (2)<sup>35</sup>

$$\sigma^2 = \frac{1}{n-1} \sum_{i=1}^n (\theta_i - \theta_0)^2 \quad (2)$$

where  $n$  is the number of bonds while  $\theta_i$  and  $\theta_0$  denote the  $i^{\text{th}}$  and ideal bond angles of the octahedron, respectively. This is shown to decrease in Fig. 3 b) as a function of increasing  $x$ , reaching zero at  $x = 0.5$  indicating ideal  $90^\circ$  O–Ti–O angles are achieved in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Although it is not fully understood, it is evident that this competition between simultaneous octahedral tilting ( $\phi$ ) and octahedral distortion ( $\sigma^2$ ) in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  leads to the compositions of  $0.2 < x < 0.4$  adopting metrically cubic unit cells. Fig. 3 c) shows that the separate A–O distances converge at  $x = 0.5$  with the average A–O distance increasing following the expansion of the A site cations and unit cell. The A–O/B–O ratio increases with  $x$  towards the value of  $\sqrt{2}$  (Table S4, ESI) which relates these distances in the cubic perovskite structure. The cubic structure of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is very similar to that of  $\text{SrTiO}_3$  with comparable lattice parameters of  $a = 3.908212(5) \text{ \AA}$  and  $3.909149(3) \text{ \AA}$ , respectively.<sup>36</sup> Refinement of ADPs to model the 18e oxygen positions in rhombohedral  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  gave similar results to those in cubic  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  where their positions were elongated normal to the central atom of the octahedron and along the direction of octahedral rotation (Fig. S1, ESI). The symmetry of space group  $R\bar{3}c$  allows for refinement of ADPs for the A site, which results in a slight elongation of the position along the  $c$  axis of the unit cell towards neighbouring octahedra (Fig. S1, ESI). Despite the static disorder generated by three different cations occupying the A site, the local deviation away from the average structure is minimal with a root mean square displacement of  $0.107(16) \text{ \AA}$  at  $300 \text{ K}$ . Most importantly, the structural analysis of the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$

solid solution and of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  in particular, shows that the  $\text{La}^{3+}$  and  $\text{K}^+$  are disordered across the twelve-coordinate A site. This disorder induced phonon-glass behaviour of the thermal conductivity in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ ,<sup>8</sup> a property expected to be replicated in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  following the observation of phonon localisation (Fig. 1).

$\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is a crystalline material that offers the high symmetry cubic structure of  $\text{SrTiO}_3$ , with the unique multi-carrier band structure suited towards thermoelectricity,<sup>13</sup> along with a disordered distribution of A

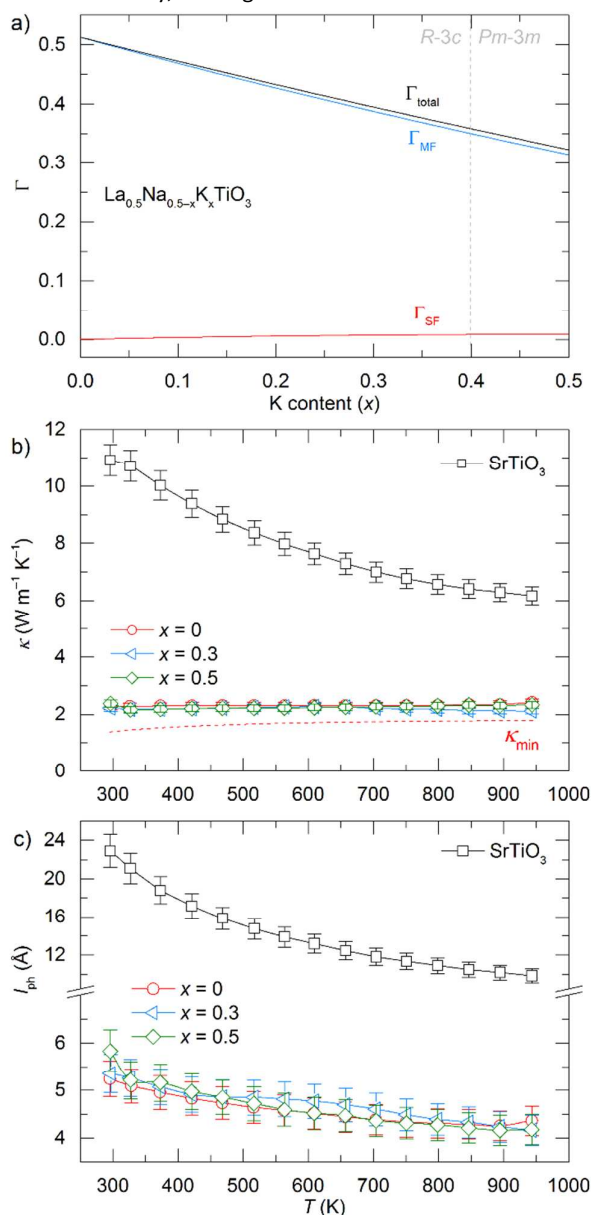


Fig. 4 Dependence of a) disorder scattering parameter ( $l$ ), including mass ( $\Gamma_{\text{MF}}$ ) and size ( $\Gamma_{\text{SF}}$ ) contributions shown in blue and red respectively, b) thermal conductivity ( $\kappa$ ), and c) phonon mean free path ( $l_{\text{ph}}$ ) on  $x$  in the solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . The theoretical minimum thermal conductivity ( $\kappa_{\text{min}}$ ) of  $\text{SrTiO}_3$  is shown as the dashed line in b) and is calculated from Cahill's model for disordered crystalline solids. [REF-Cahill1992] Phonon mean free paths in c) are obtained through eqns (4) and (5).

site metals which have considerable mass contrast. The disorder scattering parameter is defined by  $\Gamma = \Gamma_{\text{MF}} + \Gamma_{\text{SF}}$  and contributes significantly to phonon scattering in these materials. It combines contributions from the mass contrast of defects ( $\Gamma_{\text{MF}}$ ) and the size strain ( $\Gamma_{\text{SF}}$ ) induced by ionic radius variance of defects, both of which are effective disruptors of phonon transport within a crystal lattice.<sup>37</sup> The mass fluctuation and strain field terms are quantified through  $\Gamma_{\text{MF}} = \sum_i f_i (1 - m_i/\bar{m})^2$  and  $\Gamma_{\text{SF}} = \sum_i f_i (1 - r_i/\bar{r})^2$  respectively, where  $f_i$  is the fraction of atoms which have mass  $m_i$  and radius  $r_i$  occupying a crystallographic site with average mass  $\bar{m}$  and average radius  $\bar{r}$ . Fig. 4 a) shows how  $\Gamma$ ,  $\Gamma_{\text{MF}}$  and  $\Gamma_{\text{SF}}$  change as a function of  $x$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . Replacement of lighter sodium with heavier potassium results in  $\Gamma_{\text{MF}}$  decreasing from 0.513 for  $x = 0$  to 0.314 for  $x = 0.5$ , while the increased size disparity between  $\text{La}^{3+}$  and  $\text{K}^+$  raises  $\Gamma_{\text{SF}}$  from  $1.19 \times 10^{-4}$  to  $8.71 \times 10^{-3}$  for those respective values of  $x$ . Although  $\Gamma_{\text{SF}}$  is no longer negligible, with a 2.70% contribution to  $\Gamma$  in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  compared to 0.02% in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ , this increase in cation size variance is insufficient to alleviate the reduction in  $\Gamma_{\text{MF}}$  that remains the dominant contribution to the total  $\Gamma$ , which decreases from 0.513 to 0.323 between  $x = 0$  and 0.5, respectively. The measured thermal conductivities across the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  series are shown in Fig. 4 b) (Fig. S4, ESI). One of the more striking features of the experimental data is that the thermal conductivity of the compositions in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are not significantly different despite the reduction in  $\Gamma$  shown in Fig. 4 a). The compositions of  $x = 0, 0.3$  and 0.5 in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are electronically insulating meaning that  $\kappa$  is equal to  $\kappa_{\text{latt}}$  in these materials. The thermal conductivity of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  retains the phonon-glass characteristics that were observed for  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  where the temperature dependence of  $\kappa_{\text{latt}}$  matches that of the heat capacity. Additionally, both the magnitude and temperature dependence of  $\kappa_{\text{latt}}$  observed for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  are comparable with the theoretical minimum thermal conductivity ( $\kappa_{\text{min}}$ ) that would be expected if  $\text{SrTiO}_3$  exhibited phonon glass properties. This quantity, compared against experimental data in Fig. 4 b), is estimated through Cahill's model for disordered crystalline solids described by eqn (3)

$$\kappa_{\text{min}} = \left(\frac{\pi}{6}\right)^{\frac{1}{3}} k_B N^{\frac{2}{3}} \sum_i v_i \left(\frac{r_i}{\theta_i}\right)^2 \int_0^{\theta_i} \frac{x^3 e^x}{(e^x - 1)^2} dx \quad (3)$$

where  $k_B$  is the Boltzmann constant,  $N$  is the number density of atoms known from structural refinement,  $v_i$  is the velocity of sound and  $\theta_i$  is used as the cut-off frequency.<sup>38</sup> Experimental values of transverse and longitudinal velocities of sound extracted from  $\text{SrTiO}_3$  single crystals were used to calculate  $\theta_i$  through eqn (4) and are listed in Table S5 (ESI)<sup>39</sup>

$$v_s = \frac{k_B \theta_D}{h \sqrt[3]{6\pi^2 N}} \quad (4)$$

The close comparison to  $\kappa_{\text{min}}$  and glass-like dependence of  $\kappa_{\text{latt}}$  with temperature shows that  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is a crystalline material with phonon glass thermal properties. The magnitude of  $\kappa_{\text{latt}}$  at room temperature increases from 2.28 to 2.40  $\text{W m}^{-1}$

$\text{K}^{-1}$  from  $x = 0$  to 0.5 respectively, which is associated with the decrease in  $\Gamma$ . This is only a minor change in  $\kappa_{\text{latt}}$  in contrast to a considerable change to  $\Gamma$ , indicating that the disorder parameter has a smaller influence on the thermal conductivity in materials which display phonon-glass properties compared to those with phonon-crystal properties as was shown in  $\text{Sr}_{1-x}\text{La}_{0.67x}\text{TiO}_3$  where increasing  $\Gamma$  had a smaller effect as  $\kappa_{\text{latt}}$  approached  $\kappa_{\text{min}}$ .<sup>40</sup> The values of  $\kappa_{\text{latt}}$  observed for  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are very low when compared against both doped and undoped compositions of  $\text{SrTiO}_3$  which exhibit  $\kappa_{\text{latt}}$  values in the range of 4 – 12  $\text{W m}^{-1} \text{K}^{-1}$  at 300 K,<sup>41, 42</sup> and perhaps most importantly exhibit a temperature dependence that is typical of a phonon-crystal where a  $T^{-1}$  relationship is observed at high temperatures due to the anharmonic interaction of phonon-phonon Umklapp scattering processes. Data measured from undoped  $\text{SrTiO}_3$  is provided for comparison purposes in Fig. 4 b) and show that the substitution of both  $\text{La}^{3+}$  and  $\text{K}^+$  for  $\text{Sr}^{2+}$  on the A site of the perovskite structure affords a 75% reduction in thermal conductivity at 300 K (2.37(12)  $\text{W m}^{-1} \text{K}^{-1}$  for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  compared against 10.91(56)  $\text{W m}^{-1} \text{K}^{-1}$  for  $\text{SrTiO}_3$ ). This low, glass-like thermal conductivity observed experimentally for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is consistent with the results obtained from calculations (Fig. 1) which showed that phonons are localised in comparison with the phonon-crystal  $\text{SrTiO}_3$ . The calculations, performed at DFT level, show increased contributions to the phonon density of states (PDOS) in the low frequency region in comparison with  $\text{SrTiO}_3$  (Fig. S5, ESI). The largest contribution to the PDOS in  $\text{SrTiO}_3$  is at  $\approx 110 \text{ cm}^{-1}$ , however two broader contributions are observed in this region for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ ; a peak at  $\approx 95 \text{ cm}^{-1}$  predominantly consisting of contributions from La, and a second peak at  $\approx 170 \text{ cm}^{-1}$  arising from the contributions of K and Ti. Through this combined experimental and computational understanding of phonon behaviour in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ , it is clear the vibrational disorder introduced by having two high-mass contrast cations present on a single crystallographic position is highly effective at scattering phonons and reducing heat transport within the lattice, leading to the observed glass-like thermal conductivity in crystalline  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . This increased localisation of phonons is similar to that observed in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ ,<sup>8</sup> and shows that the disordered  $\text{La}^{3+}$  and  $\text{K}^+$  cations in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  are responsible for the phonon-glass behaviour by acting as decoupled local oscillators with incoherent phases which dramatically reduce time and distance between phonon scattering events. Experimentally, this is evident from the phonon mean free paths that are shown in Fig. 4 c) and are calculated from eqns (4) and (5) using  $\theta_D$  of 630 K.<sup>43</sup>

$$\kappa_{\text{latt}} = \frac{1}{3} \cdot C_V \cdot l_{\text{ph}} \cdot v_s \quad (5)$$

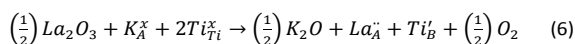
where  $C_V$  is the isochoric heat capacity, and  $l_{\text{ph}}$  is the phonon mean free path. The phonon mean free paths of compositions in the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  series all show relatively flat temperature dependence in comparison with  $\text{SrTiO}_3$  which is a clear indicator of the emergence of glassy characteristics in these materials. The magnitudes of  $l_{\text{ph}}$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  of 4-

6 Å are comparable to those of vitreous materials such as SiO<sub>2</sub> which typically are around 5–7 Å.<sup>44</sup> Microstructure was shown to influence  $\kappa_{\text{latt}}$  and  $I_{\text{ph}}$  of SrTiO<sub>3</sub>,<sup>16</sup> however, this required nanometre-sized grains. The effect of microstructure in La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> materials is negligible due to the large disparity between  $I_{\text{ph}}$  and grain size, which are shown through SEM (Fig. S6, ESI) to be much larger and of the order of 10–20 µm.

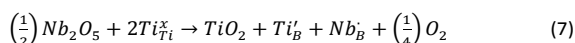
#### Introduction of electronic charge carriers into La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub>

One of the more useful features resulting from the complete replacement of Sr<sup>2+</sup> for a charge balanced mixture of La<sup>3+</sup> and K<sup>+</sup> in La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> is that the chemical and structural flexibility of the perovskite structure is maintained. Electron doping in SrTiO<sub>3</sub> involves the incorporation of an aliovalent cation with higher charge, such as Ln<sup>3+</sup> on the A site or Nb<sup>5+</sup> on the B site,<sup>42, 45, 46</sup> which through charge compensation mechanisms reduces a sufficient amount of Ti<sup>4+</sup> to Ti<sup>3+</sup> giving rise to electronic transport. Even though the A site in La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> has a disordered cation arrangement, the flexible structural chemistry enables the introduction of electronic carriers through doping of either the A or B site. Incorporation of Nb<sup>5+</sup> onto the B site through La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> allows for the mechanisms that govern the thermal and electronic properties to be confined to the separate A and B sites of the perovskite structure, respectively, as was reported for PGEC La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>1–x</sub>Nb<sub>x</sub>O<sub>3</sub>.<sup>8</sup> Doping of the A site can be achieved by targeting non-stoichiometric compositions through La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> (with  $y < 0.5$ ). This maintains the mass contrast that affords the phonon-glass thermal conductivity, and usefully allows for the same nominal charge carrier concentrations as La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> to be accessed by using smaller dopant amounts due to the nature of the dopant mechanisms which are described by Kröger-Vink notation in reaction eqns (6) and (7), where A and B refer to the site occupied.

For La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub>:



For La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub>:



Through solid state reaction performed at 1300 °C under a H<sub>2</sub>/N<sub>2</sub> (5%) atmosphere, it is possible to produce phase pure samples of La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> with  $0.4 \leq y \leq 0.5$  and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> up to  $z = 0.1$ . These values define the limits of doping for each material under these particular synthetic conditions; decreasing  $y$  below 0.4 in La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> results in the formation of La<sub>2</sub>Ti<sub>2</sub>O<sub>7</sub> impurities, while increasing  $z$  further than 0.1 yields LaNbO<sub>4</sub> as an impurity phase in La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub>. The effects of A and B site doping upon the unit cell parameters of La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> are shown in Figs. 5 a) and b), respectively. A linear decrease in unit cell volume is observed as a function of La<sup>3+</sup> content in La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> associated with the larger K<sup>+</sup> cation being replaced with La<sup>3+</sup> ( $r_{\text{K}^+} = 1.64$  Å and  $r_{\text{La}^{3+}} = 1.36$  Å).<sup>20</sup> Both La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> compositions are indexed to rhombohedral  $R\bar{3}c$  symmetry with separation of primitive cell parameters  $a_p$  and  $c_p$ , shown in Fig. 5 a) (Tables S6, S7 and S8,

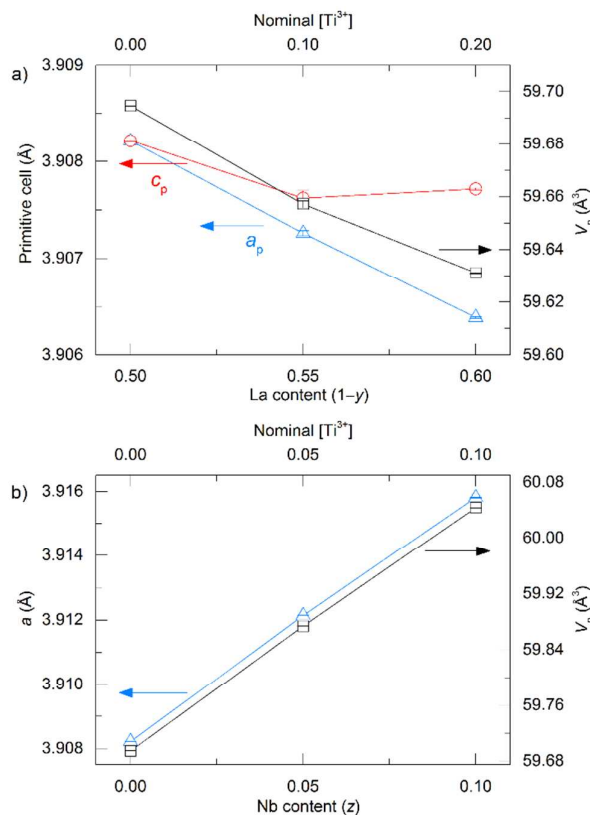


Fig. 5 Variation in a) primitive unit cell parameters and volume as a function of La<sup>3+</sup> content in A site doped La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> with  $R\bar{3}c$  symmetry, and b) cubic unit cell parameter and volume in B site doped La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> ( $Pm\bar{3}m$ ) as a function Nb<sup>5+</sup> substitution. The nominal Ti<sup>3+</sup> concentrations are plotted on the top axes of each and increase with increasing La<sup>3+</sup> and Nb<sup>5+</sup> content. The La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> series are cubic ( $Pm\bar{3}m$ ), while La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> materials are rhombohedral ( $R\bar{3}c$ ).

Figs. S7 and S8, ESI). This distortion away from the cubic symmetry of the La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> parent results from the decreasing average cation size on the A site, as well as an increasing cation size on the B site from the reduction of Ti<sup>4+</sup> to Ti<sup>3+</sup> ( $r_{\text{Ti}^{4+}} = 0.605$  Å and  $r_{\text{Ti}^{3+}} = 0.67$  Å); the tolerance factor can be reduced by decreasing  $r_A$  or increasing  $r_B$ . Contrary to this, the cubic symmetry of the La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> parent is maintained in the B site doped La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> materials up to  $z = 0.1$  with a linear increase in the unit cell parameter and volume shown in Fig. 5 b) (Tables S9, S10 and S11, Figs. S9 and S10, ESI), consistent with the substitution of Ti<sup>4+</sup> with larger Nb<sup>5+</sup> ( $r_{\text{Ti}^{4+}} = 0.605$  Å and  $r_{\text{Nb}^{5+}} = 0.64$  Å). There remains no evidence for ordering of the A site cations even in non-stoichiometric La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub>, indicating that the disordered distributions of La<sup>3+</sup> and K<sup>+</sup>, which we have shown to be essential for the phonon glass  $\kappa$  in these materials, are preserved in all materials reported in this study. The refined oxygen occupancies for the A and B doped materials studied through both SXRD and NPD did not provide any evidence for oxide deficiencies (Tables S7 and S10,

Nominal composition	La	K	Ti	Nb
La <sub>0.5</sub> K <sub>0.5</sub> TiO <sub>3</sub>	0.518(3)	0.507(5)	1.000(5)	–
La <sub>0.55</sub> K <sub>0.45</sub> TiO <sub>3</sub>	0.556(3)	0.445(2)	1.000(5)	–
La <sub>0.6</sub> K <sub>0.4</sub> TiO <sub>3</sub>	0.608(3)	0.413(2)	1.000(7)	–
La <sub>0.5</sub> K <sub>0.5</sub> Ti <sub>0.95</sub> Nb <sub>0.05</sub> O <sub>3</sub>	0.505(2)	0.493(4)	0.950(3)	0.0487(1)
La <sub>0.5</sub> K <sub>0.5</sub> Ti <sub>0.9</sub> Nb <sub>0.1</sub> O <sub>3</sub>	0.507(7)	0.469(2)	0.900(6)	0.0981(1)



ESI). When refined, values within error of a fully occupied oxygen position were obtained much like the oxygen occupancies in the undoped materials  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  and  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  (Tables S1 and S2, ESI).

The metal contents for each of the doped materials were determined using ICP-OES and are listed in Table 2. Both the La/K and Ti/Nb ratios agree very well with those from the nominal compositions, and compare closely to the refined occupancies obtained from Rietveld analysis of SXRD and PND data (Tables S1, S6, S7, S9 and S10, ESI). The analysis shows near-stoichiometric ratios for La and K in the undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  parent, in contrast to the single previous report of cubic  $\text{La}_{0.531}\text{K}_{0.469}\text{TiO}_3$ , in which not all of the alkali metal was incorporated into the structure.<sup>23</sup>

Table 2 Results of ICP-OES measurements on  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  materials. Values are normalised to titanium and compared against the nominal expected compositions. Metal contents were corrected by analysing metal solutions of known concentrations both separately and combined. Correction factors are La = 1.01, K = 1.10, Ti = 1.00 and Nb = 1.00. Absolute errors are shown.

The sintering of cold-pressed pellets of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions at 1350 °C under a  $\text{H}_2/\text{N}_2$  (5%) atmosphere yielded ceramics with measured densities higher than 90% of the values expected from structural analysis (Table S12, ESI), and SEM imaging showed the pellets to have minimal porosity (Fig. S6, ESI).

The temperature dependences of the thermal conductivity for doped  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions are shown in Fig. 6 a). These materials exhibit low phonon-glass thermal conductivities comparable to that of undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  despite possessing electron carrier concentrations, determined from nominal compositions, which lie within the range expected for heavily doped semiconductors ( $10^{19}$ - $10^{21} \text{ cm}^{-3}$ ). The thermal conductivities have magnitudes of 2.45(12)-2.65(13)  $\text{W m}^{-1} \text{ K}^{-1}$  at room temperature and 2.45(12)-2.80(15)  $\text{W m}^{-1} \text{ K}^{-1}$  at 850 K and have a temperature dependence like  $\kappa_{\text{min}}$  determined from eqn (3). In contrast to the thermal conductivities of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  which show no correlation with doping, the electronic conductivities and Seebeck coefficients display a strong doping dependence shown in Figs. 6 b) and c). In  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ , peak conductivities increase from 180(9) to 212(10)  $\text{S cm}^{-1}$  at 600 K for  $z = 0.05$  and 0.1 respectively, whereas the peak values of  $\sigma$  in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  increase from

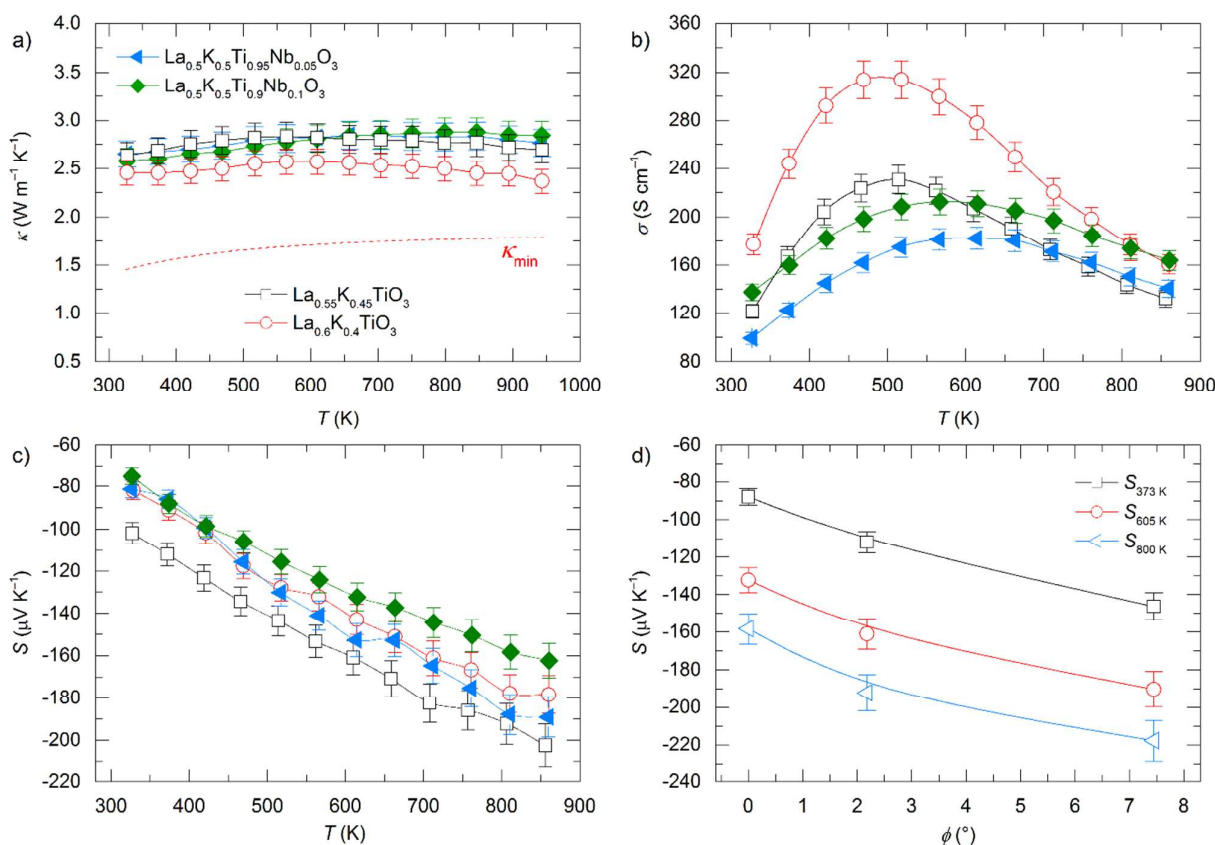


Fig. 6 Temperature dependence of a) thermal conductivity ( $\kappa$ ), b) electronic conductivity ( $\sigma$ ), and c) Seebeck coefficient ( $S$ ) for A site ( $\text{La}_{1-y}\text{K}_y\text{TiO}_3$ ,  $y = 0.45$  and  $0.4$ ) and B site ( $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ ,  $z = 0.05$  and  $0.1$ ) doped compositions of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Dependence of  $S$  as a function of tilt angle ( $\phi$ ) at different temperatures is shown in d) using three materials that have the same nominal carrier concentration:  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  ( $\phi = 0^\circ$ ) and  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  ( $\phi = 2.18(1)^\circ$ ) from this study, and  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  ( $\phi = 7.45(3)^\circ$ ) which was taken from a recent study of PGEC  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  materials with  $R3c$  symmetry.[REF-Daniels2017]

230(11) S cm<sup>-1</sup> for  $y = 0.45$  to 315(16) S cm<sup>-1</sup> for  $y = 0.4$  at 500 K, associated with increasing carrier concentration. The observed temperature dependencies of both  $\sigma$  and  $S$  for all compositions are similar to many doped titanate thermoelectrics reported in the literature.<sup>40, 42, 46, 47</sup> Maxima in  $\sigma$  are observed between 500 – 600 K and are associated with increased resistance from acceptor-based phases which are localised to the grain boundaries.<sup>48</sup> The negative values of  $S$  indicate electron doping of predominantly  $n$ -type carriers within all compositions, and both A and B site doped materials exhibit decreasing  $|S|$  with increasing carrier concentration which is expected for metals or parabolic band semiconductors through eqn (8)

$$S = \frac{8\pi^2 k_B^2}{3eh^2} m^* T \left( \frac{\pi}{3n} \right)^{2/3} \quad (8)$$

where  $h$  is Planck's constant,  $e$  is elementary charge,  $m^*$  is the carrier effective mass and  $n$  is the carrier concentration.<sup>3</sup> For La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub>, which nominally have the same  $n$ , almost identical conductivities of  $\approx 210(10)$  S cm<sup>-1</sup> are observed at 600 K. Interestingly for these particular compositions,  $|S|$  is considerably higher for La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> than for La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub>, with values of  $-161(8)$  and  $-132(7)$   $\mu$ V K<sup>-1</sup> at 600 K respectively, and  $|S|$  of La<sub>0.6</sub>K<sub>0.4</sub>TiO<sub>3</sub> is still higher than that of La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> despite having a higher nominal carrier concentration. As La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> both have the same  $n$  but a larger  $|S|$  is observed experimentally for La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub>, from the relationship shown in eqn (8) this indicates that a higher  $m^*$  is achieved through the A site doping of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> compared to B site doping through La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. This shows how the slight differences in the crystal structures of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub> can significantly influence the electronic properties. In general, 3d-2p orbital overlap in these perovskites is maximised through shorter bonding and/or tilt angles that approach 0°. In La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub> the Nb<sup>5+</sup> substitution retains the linear Ti–O–Ti bonding of the cubic ( $Pm\bar{3}m$ ) structure (Table S11, ESI). This maximises band dispersion and therefore  $\sigma$ , but leads to a lower  $m^*$  which supports the decreased  $|S|$  observed experimentally for La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. In contrast, the change of symmetry to  $R\bar{3}c$  in La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> reduces the orbital overlap through octahedral tilting which reduces band dispersion yielding a larger  $m^*$  and is the reason that higher  $|S|$  values are observed for La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> compared to La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. The decrease in  $\sigma$  that a larger  $m^*$  is expected to bring about is offset by the increased overlap between neighbouring Ti 3d-t<sub>2g</sub> orbitals in the smaller, more distorted cell of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> which accounts

for the closely matched conductivities observed for La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub>.<sup>42, 50</sup> Using the additional data available from the previous report of La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>1-x</sub>Nb<sub>x</sub>O<sub>3</sub>,<sup>8</sup> it is possible to observe the relationship between  $|S|$  and octahedral tilt angle, which is plotted in Fig. 6 d) for compositions with the same nominal carrier concentration and further highlights the effects of tilting on band dispersions discussed above. La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> is cubic ( $Pm\bar{3}m$ ) with zero tilt, while both La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> are rhombohedral ( $R\bar{3}c$ ) with tilt angles of 2.18(1)° and 7.45(3)° respectively. The magnitude of  $|S|$  increases with increasing tilt angle at all temperatures with a  $\approx 60$   $\mu$ V K<sup>-1</sup> difference between La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> and La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> due to increased  $m^*$  as a result of reduced orbital overlap brought about by octahedral tilting. In contrast to the clear relationship between  $|S|$  and the tilt angle, similar magnitudes of  $\sigma$  are observed for these materials with the same nominal carrier concentration. From this it is clear that despite targeting high carrier mobility by tuning the symmetry of the unit cell to maximise band dispersions in the zero-tilt structure of La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub>, the optimisation of  $\sigma$  is far more complex and depends significantly on the processing conditions of the measured ceramics with extrinsic factors such as grain boundary resistance ultimately dominating  $\sigma$ .<sup>42, 48</sup> The Seebeck coefficient, measured at zero current transport, should be insensitive to grain boundary effects and exhibits a clear correlation to the octahedral tilt angle.

As a result of the improved  $|S|$  values observed for La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> significant differences are observed between the power factors of the A and B site doped materials as shown in Fig. 7 a). Enhancements in  $\sigma$ , particularly at low temperatures, alongside the increased  $m^*$  and  $|S|$  in La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> result in a power factors of 0.58(5) mW m<sup>-1</sup> K<sup>-2</sup> at 600 K compared to 0.39(3) mW m<sup>-1</sup> K<sup>-2</sup> in La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. The observation of improved power factors in titanate perovskite thermoelectrics with smaller, more-distorted unit cells is commonplace;  $S^2\sigma$  was enhanced with increasing values of  $x$  in the solid solution Ba<sub>0.9-x</sub>Sr<sub>x</sub>La<sub>0.1</sub>TiO<sub>3</sub>,<sup>51</sup> and the largest  $S^2\sigma$  values were achieved for the smallest lanthanides studied in Sr<sub>0.9</sub>R<sub>0.1</sub>TiO<sub>3- $\delta$</sub> .<sup>42</sup> In this study, when considering both  $\sigma$  and  $S$ , the electronic transport is improved in La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> compositions ( $y = 0.45$  and  $0.4$ ) across the entire measured temperature range compared to B site doped La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub> as shown in Fig. 7 a) and is ascribed to the improved Seebeck coefficients which result from lower band dispersions and higher  $m^*$  in the more distorted structures of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub>. The power factors of the La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> series are comparable to those of La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>1-x</sub>Nb<sub>x</sub>O<sub>3</sub> reported previously.<sup>8</sup>

Even though higher power factors are achieved for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  compared against  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ , both series retain the same glass-like thermal conductivity of the undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  parent and the electronic contribution to the thermal conductivity ( $\kappa_{\text{elec}}$ ) is no larger than 12% for all samples (Fig. S11, ESI). This highlights the weak correlation between electronic and thermal transport within these materials, properties that are usually very strongly coupled in titanate thermoelectrics with  $\kappa_{\text{elec}}$  contributing  $\approx 25\%$  of the total thermal conductivity in similarly doped  $\text{Sr}_{1-x}\text{La}_x\text{TiO}_{3-\delta}$ ,<sup>15</sup> and again demonstrates the advantage of utilising cation mass disorder to control  $\kappa_{\text{latt}}$  which remains the dominant contributor to  $\kappa$  in these materials.

Peak  $ZT$  values of  $\approx 0.18$  are observed at 800 K in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  compositions and the figures of merit increase with temperature for all materials, as shown in Fig. 7 b). These are slightly lower than the highest values obtained previously from  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  and result from the combination of comparable power factors and slightly higher thermal conductivities in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ . The improved performance of A site doped  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  over B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  is clear, not only in peak  $ZT$  but also for the entire measurement range. This increased figure of merit is attributed to the enhanced power factors (Fig. 7a)) which originate from decreased band dispersions in more distorted  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  materials. At all temperatures, the  $ZT$  values of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  are typically 20-50% higher than those of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  with the largest differences in the intermediate temperature range (350-700 K). As a result of this, the average energy conversion efficiencies ( $\varepsilon_{\text{ave}}$ ) of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  are a respectable 1.39-1.59%, for  $y = 0.45$  and  $0.4$  respectively, compared against 0.093-0.097% for in  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  respectively, when evaluated through eqn (9) over the temperature ranges applicable to the recovery of waste heat from automotive exhausts.<sup>4</sup>

$$\varepsilon_{\text{ave}} = \frac{\frac{T_H - T_C}{T_H} \int_{T_C}^{T_H} \frac{\sqrt{ZT+1}-1}{\sqrt{ZT+1} + (T_C/T_H)} dT}{T_H - T_C} \quad (9)$$

The large difference in efficiencies between  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  is due to the increased figures of merit across the temperature range 350-700 K, and not due to the highest peak value. The importance of  $ZT_{\text{average}}$ , particularly for application temperature ranges, is frequently overlooked when searching for new peak  $ZT$  materials.

## Conclusions

This work presents the study of a new solid solution between phonon-glass  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  ( $R\bar{3}c$ ) and  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  ( $Pm\bar{3}m$ ). It is possible to tune the cation size on the A site in order to target the cubic structure with linear Ti–O–Ti bonding that is associated with the high electronic conductivity of  $\text{SrTiO}_3$  based thermoelectrics.  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is the first reported cubic phonon-glass titanate thermoelectric. The phonon-glass thermal conductivity results from localised phonon interactions caused by the disordered distribution of high mass

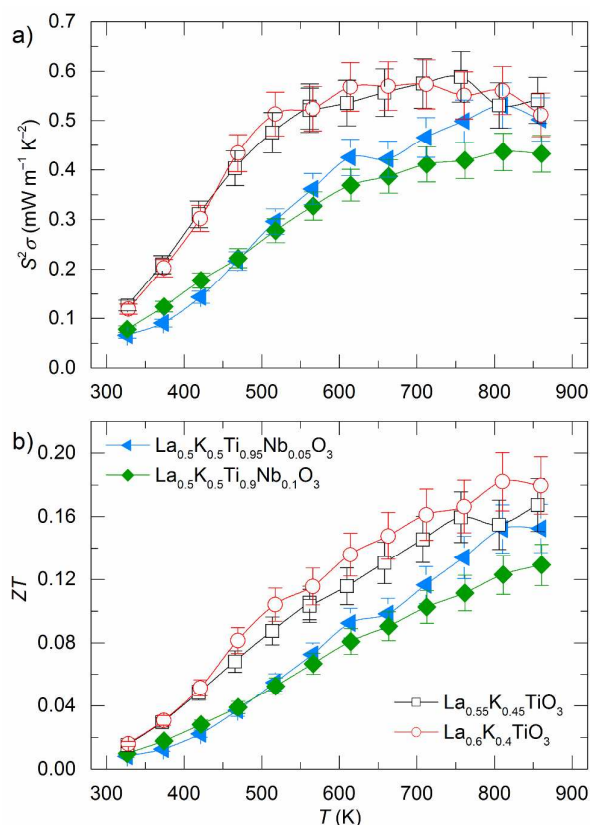


Fig. 7 Temperature dependence of a) thermoelectric power factors ( $S^2\sigma$ ) and b) thermoelectric figures of merit ( $ZT$ ) for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions

contrast  $\text{La}^{3+}$  and  $\text{K}^+$ . Electronic carriers can be introduced through either A or B site doping in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ , respectively. The A site doping through  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  leads to larger power factors and a higher overall  $ZT$  performance compared to B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ . We find that the electronic conductivity is dominated by extrinsic effects such as grain boundary resistance despite band dispersions being optimised through the cubic symmetry of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The Seebeck coefficient is unaffected by the extrinsic factors and this performance gain is associated with the more distorted and lower symmetry structures of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  ( $R\bar{3}c$ ) compositions which decrease band dispersion and lead to higher effective masses and Seebeck coefficients compared with  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  ( $Pm\bar{3}m$ ). Phonon-glass thermal conductivities are retained for both A and B site doped materials providing a further example of how intrinsic cation mass disorder can minimise the correlation between the usually strongly coupled thermal and electronic transport in perovskites and yield PGEC behaviour in a cubic titanate thermoelectric. The control of power factors through the distortion of the unit cell could provide guidance in the choice of future target materials, such as combining high mass contrast compositions with smaller cations to target improved electronic transport (e.g.  $\text{Ln}_{0.5}\text{A}_{0.5}\text{TiO}_3$  where  $A = \text{Na}^+/\text{K}^+$ , and  $\text{Ln} = \text{Dy}^{3+}/\text{Ho}^{3+}$ ).

## Conflicts of interest

There are no conflicts to declare.

## Acknowledgements

We thank EPSRC for funding under EP/N004884. We thank the Materials Chemistry Consortium (EPSRC: EP/L000202) for provision of computer time on ARCHER UK National Supercomputing Service. We thank Diamond Light Source and ISIS for provision of beamtime, and Claire Murray, Sarah Day and Paul Henry for assistance on the I11 and HRPD instruments. **Marco Zanella is thanked for measurement of SEM images.**

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**A and B site doping of a phonon-glass perovskite oxide thermoelectric**

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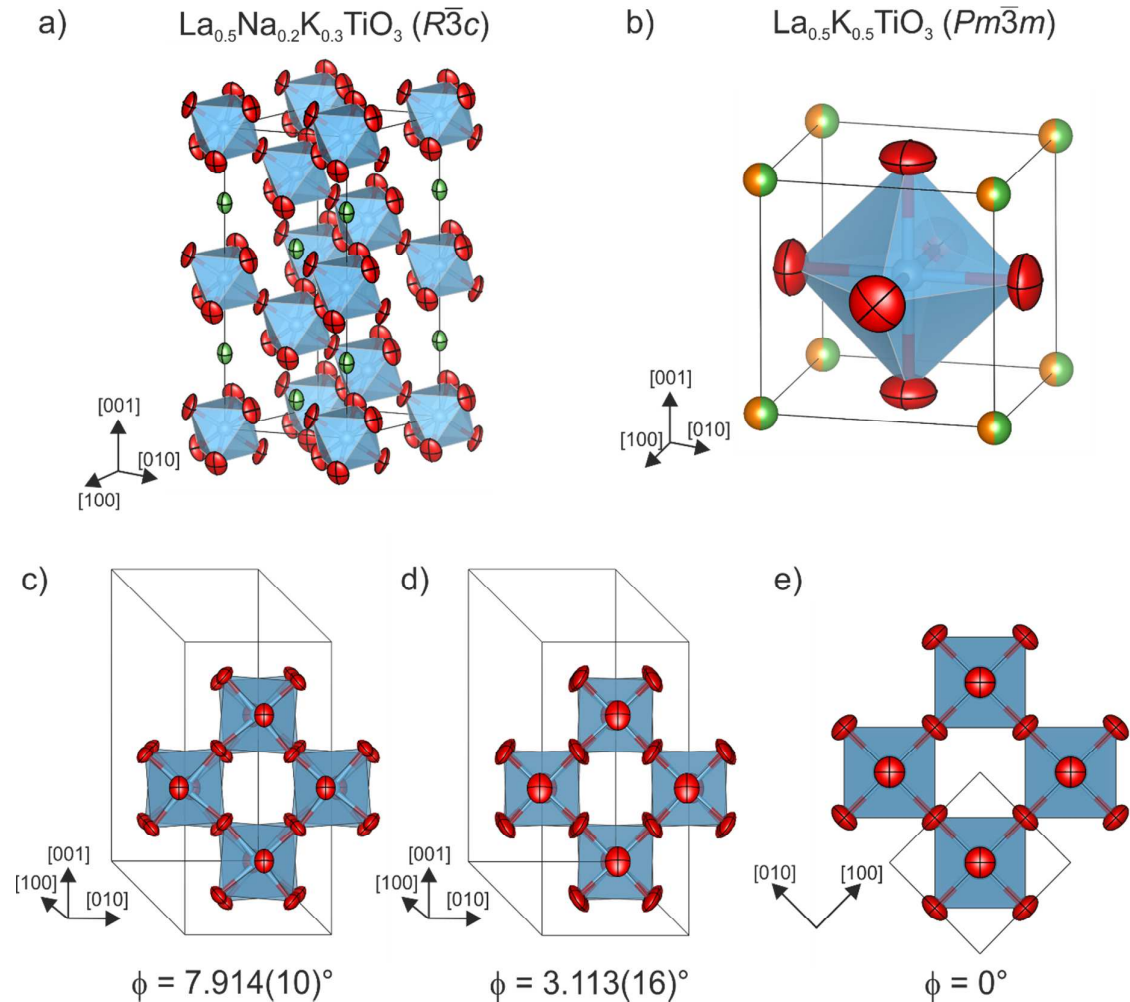
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**Table S1.** Structural parameters of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). O sites are modelled using anisotropic displacement parameters. Space group =  $Pm\bar{3}m$  with dimensions  $a = 3.908212(5) \text{ \AA}$ , and calculated density =  $5.09091(2) \text{ g cm}^{-3}$ . Refinement statistics are  $R_{\text{wp}} = 1.485$ ,  $R_{\text{exp}} = 0.623$ ,  $\chi^2 = 5.667$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	1a	0	0	0	0.4817(8)	0.704(4)
K	1a	0	0	0	0.5183(8)	0.704(4)
Ti	1b	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	1.00	0.644(6)
O	3c	0	$\frac{1}{2}$	$\frac{1}{2}$	0.9994(9)	1.446(16)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
O	0.0066(2)	0.01842(13)	0.01842(13)	0	0	0

\* Symmetry conditions for O site:  $U_{11}, U_{22}=U_{33}, U_{12}=U_{13}=U_{23}=0$



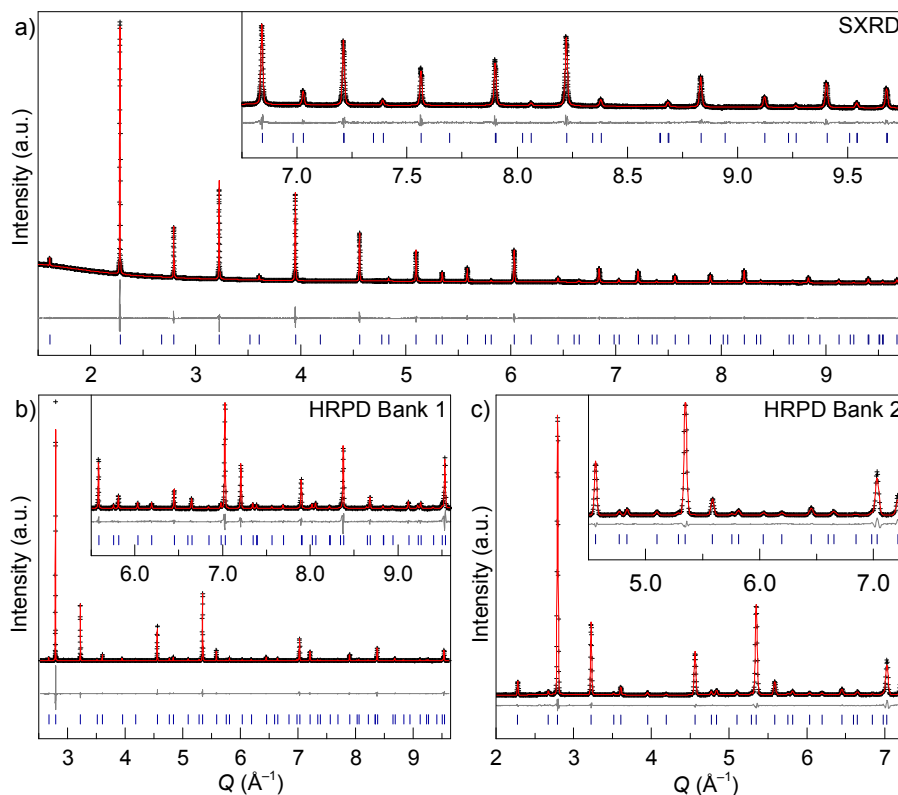
**Fig. S1.** The rhombohedral ( $R\bar{3}c$ ) unit cell of  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  is shown in a), and the cubic ( $Pm\bar{3}m$ ) unit cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is shown in b). A sites are coloured green/yellow, B sites coloured blue and oxygen positions in red. Atom sizes are set by  $U_{\text{iso}}$ . Octahedral tilting in  $x = 0, 0.3$  and  $0.5$  compositions of  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are shown in c), d) and e), respectively.

**Table S2.** Structural parameters of  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). Coherent neutron scattering lengths: La = 8.24 fm, Na = 3.63 fm, K = 3.67 fm, Ti = -3.438 fm, and O = 5.803 fm.<sup>1</sup> La/Na/K and O sites are modelled using anisotropic displacement parameters. Space group =  $R\bar{3}c$  with dimensions  $a = b = 5.50915(11)$  Å,  $c = 13.4941(6)$  Å,  $\gamma = 120^\circ$ , and calculated density =  $5.0755(2)$  g cm<sup>-3</sup>. Refinement statistics are  $R_{\text{wp}} = 3.004$ ,  $R_{\text{exp}} = 0.623$ ,  $\chi^2 = 23.218$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å <sup>2</sup> ]
La	6a	0	0	1/4	0.493(1)	1.15(3)
Na	6a	0	0	1/4	0.198(7)	1.15(3)
K	6a	0	0	1/4	0.302(7)	1.15(3)
Ti	6b	0	0	0	1.00	0.505(10)
O	18e	0.51564(12)	0	1/4	0.999(5)	1.656(16)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
La/Na/K	0.00860(14)	0.00860(14)	0.0172(6)	0.00430(7)	0	0
O	0.0161(8)	0.0082(2)	0.0254(3)	0.0041(8)	-0.0038(4)	-0.0076(9)

\* Symmetry conditions for La/Na/K site:  $U_{11}=U_{22}=2U_{12}$ ,  $U_{33}$ , and  $U_{13}=U_{23}=0$

\*\* Symmetry conditions for O site:  $U_{11}$ ,  $U_{22}=2U_{12}$ ,  $U_{33}$ , and  $2U_{13}=U_{23}$

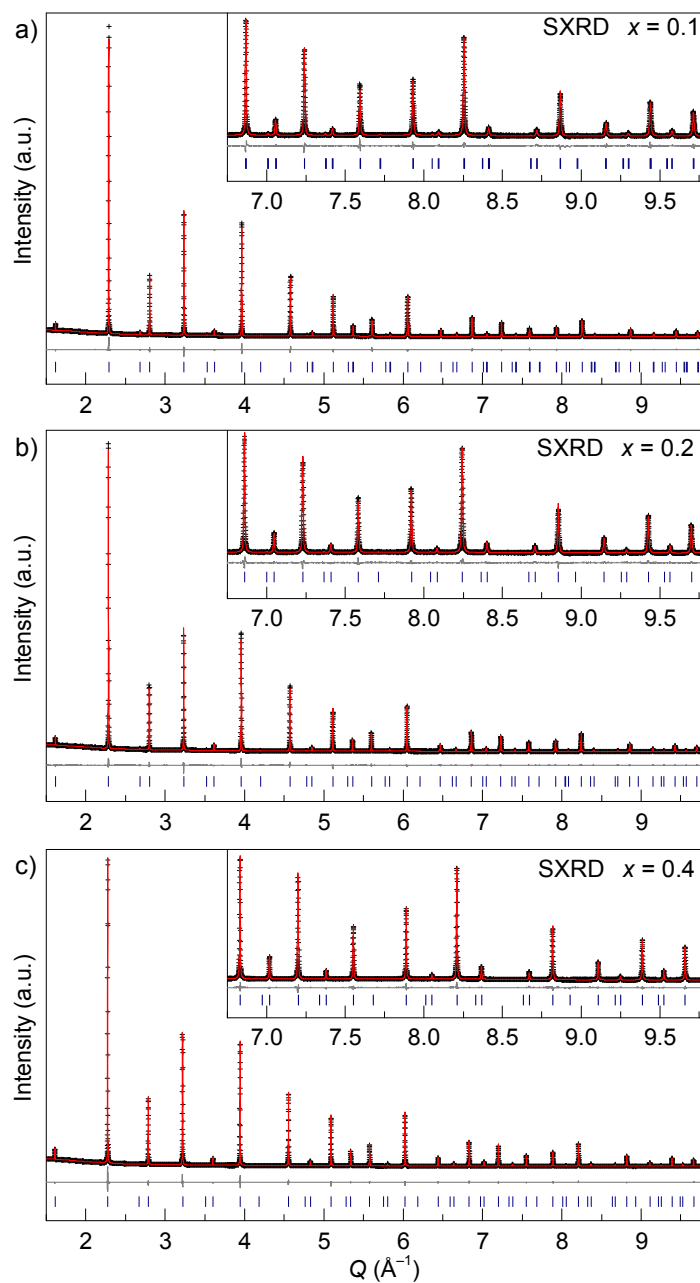


**Fig. S2.** Rietveld refinements against a) SXRD ( $\lambda = 0.82608$  Å) and high resolution neutron powder diffraction (NPD) data measured on b) Bank 1 (backscattering  $2\theta = 168^\circ$ ) and c) Bank 2 ( $2\theta = 90^\circ$  bank) of HRPD at room temperature for  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  using space group  $R\bar{3}c$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks.

**Table S3.** Structural parameters of  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  ( $x = 0.1, 0.2$  and  $0.4$ ) series from Rietveld refinement against SXRD data only (298 K).

$x = 0.1, \text{La}_{0.5}\text{Na}_{0.4}\text{K}_{0.1}\text{TiO}_3$						
Space group = $R\bar{3}c$ , $a = b = 5.489273(11) \text{ \AA}$ , $c = 13.43819(6) \text{ \AA}$ , $\gamma = 120^\circ$ , and $\rho_{\text{calc}} = 5.06335(2) \text{ g cm}^{-3}$ , refinement statistics are $R_{\text{wp}} = 2.543$ , $R_{\text{exp}} = 0.941$ , $\chi^2 = 7.311$						
Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	6a	0	0	$\frac{1}{4}$	0.5	0.363(3)
Na	6a	0	0	$\frac{1}{4}$	0.413(3)	0.363(3)
K	6a	0	0	$\frac{1}{4}$	0.087(3)	0.363(3)
Ti	6b	0	0	0	1.0	0.299(4)
O	18e	0.53364(12)	0	$\frac{1}{4}$	1.0	0.759(12)
$x = 0.2, \text{La}_{0.5}\text{Na}_{0.3}\text{K}_{0.2}\text{TiO}_3$						
Space group = $R\bar{3}c$ , $a = b = 5.49564(9) \text{ \AA}$ , $c = 13.4613(5) \text{ \AA}$ , $\gamma = 120^\circ$ , and $\rho_{\text{calc}} = 5.0798(3) \text{ g cm}^{-3}$ , refinement statistics are $R_{\text{wp}} = 2.639$ , $R_{\text{exp}} = 1.109$ , $\chi^2 = 5.675$						
Atom	Wyckoff site	x	Y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	6a	0	0	$\frac{1}{4}$	0.5	0.326(3)
Na	6a	0	0	$\frac{1}{4}$	0.331(3)	0.326(3)
K	6a	0	0	$\frac{1}{4}$	0.169(3)	0.326(3)
Ti	6b	0	0	0	1.0	0.328(4)
O	18e	0.52412(17)	0	$\frac{1}{4}$	1.0	1.103(12)
$x = 0.4, \text{La}_{0.5}\text{Na}_{0.1}\text{K}_{0.4}\text{TiO}_3$						
Space group = $R\bar{3}c$ , $a = b = 5.51816(6) \text{ \AA}$ , $c = 13.5170(3) \text{ \AA}$ , $\gamma = 120^\circ$ , and $\rho_{\text{calc}} = 5.10288(15) \text{ g cm}^{-3}$ , refinement statistics are $R_{\text{wp}} = 2.020$ , $R_{\text{exp}} = 0.850$ , $\chi^2 = 5.641$						
Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	6a	0	0	$\frac{1}{4}$	0.5	0.658(2)
Na	6a	0	0	$\frac{1}{4}$	0.143(3)	0.658(2)
K	6a	0	0	$\frac{1}{4}$	0.357(3)	0.658(2)
Ti	6b	0	0	0	1.0	0.565(3)
O	18e	0.5069(4)	0	$\frac{1}{4}$	1.0	1.325(12)





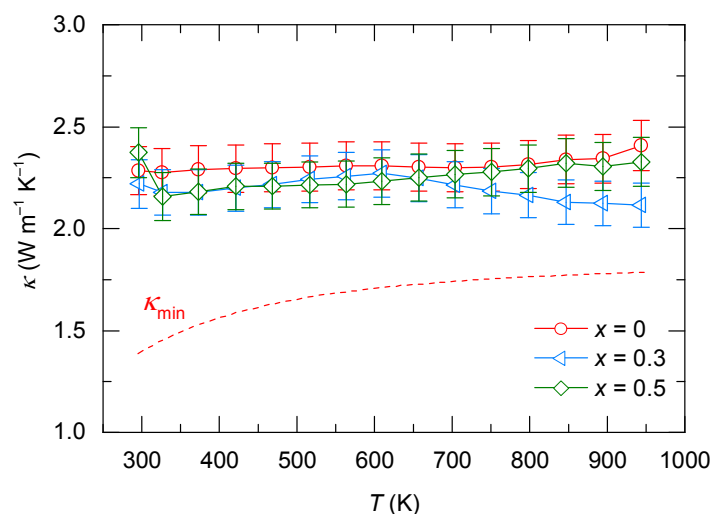
**Fig. S3.** Rietveld refinements made against synchrotron X-ray diffraction (SXR) data ( $\lambda = 0.82608 \text{ \AA}$ ) for a)  $x = 0.1$ , b)  $x = 0.2$  and c)  $x = 0.4$  compositions in the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  solid solution. Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks.

**Table S4.** Bond distances and angles as a function of  $K^+$  substitution in the  $La_{0.5}Na_{0.5-x}K_xTiO_3$  series obtained from Rietveld refinement against room temperature SXRD and high resolution NPD data.

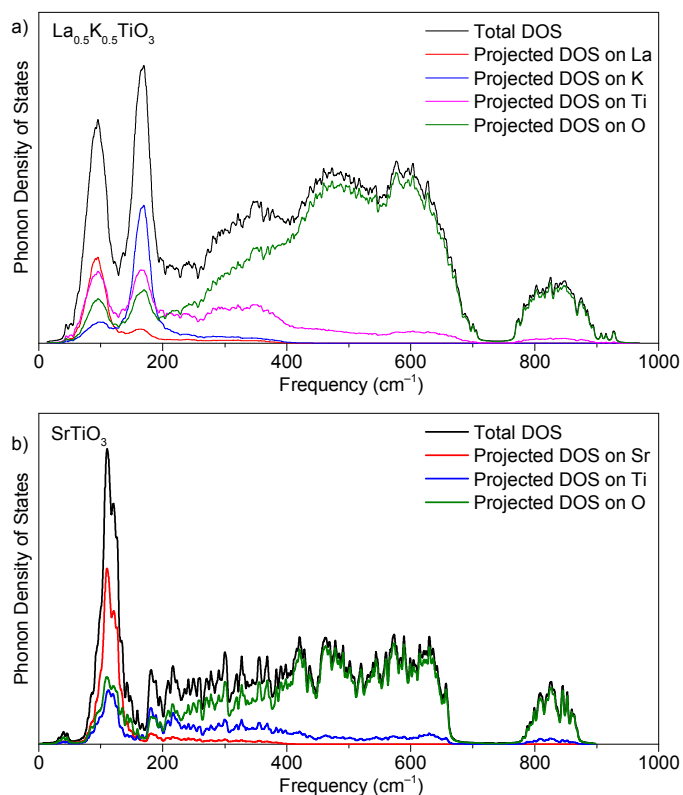
$x$	A–O [Å]	Average A–O [Å]	B–O [Å]	A–O/B– O	B–O–B [°]	$\phi$ [°]
0.0	$3 \times 2.5213(3),$ $6 \times 2.74784(3),$ $3 \times 2.9613(3)$	2.74457(17)	1.95003(4)	1.4075	167.044(14)	7.914(10)
0.1	$3 \times 2.5600(6),$ $6 \times 2.74979(6),$ $3 \times 2.9293(6)$	2.7472(3)	1.94915(8)	1.4095	167.13(4)	6.65(2)
0.2	$3 \times 2.6153(9),$ $6 \times 2.75106(6),$ $3 \times 2.8805(9)$	2.7495(5)	1.94757(9)	1.4118	172.19(5)	4.78(3)
0.3	$3 \times 2.6680(4),$ $6 \times 2.75591(2),$ $3 \times 2.8409(4)$	2.7552(2)	1.94965(3)	1.4132	174.92(2)	3.113(16)
0.4	$3 \times 2.721(2),$ $6 \times 2.75939(7),$ $3 \times 2.797(2)$	2.7592(10)	1.95135(10)	1.4140	177.76(13)	1.37(8)
0.5	$12 \times 2.76352$	2.76352	1.95411	1.4142	180	0

**Table S5.** Parameters used in determination of theoretical minimum thermal conductivity ( $\kappa_{min}$ ) of  $SrTiO_3$  through the Cahill model for disordered crystalline materials.<sup>2</sup> Values of transverse ( $v_{trans}$ ) and longitudinal ( $v_{long}$ ) sound velocities were extracted from single crystals and taken from the literature.<sup>3</sup> These were used to calculate the cut-off frequencies ( $\theta_i$ ) from both through equation (3) in the main text which were then summed as separate contributions to  $\kappa_{min}$  in equation (2).

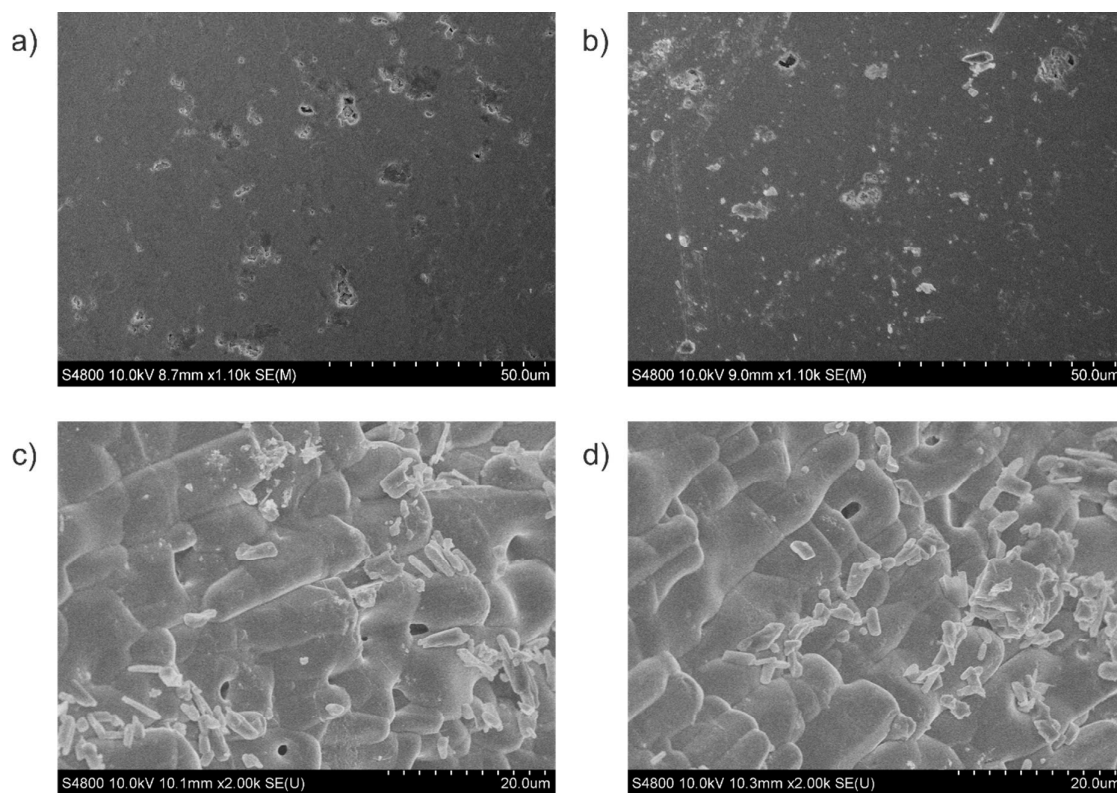
Material	$v_{trans}$ [ $m s^{-1}$ ]	$\theta_{trans}$ [K]	$v_{long}$ [ $m s^{-1}$ ]	$\theta_{long}$ [K]	$\kappa_{300 K}$ [ $W m^{-1} K^{-1}$ ]	$K_{700 K}$ [ $W m^{-1} K^{-1}$ ]
$SrTiO_3$	4705	612.25	7900	1028.02	1.398	1.741



**Fig. S4.** Thermal conductivity ( $\kappa$ ) for  $x = 0, 0.3$  and  $0.5$  in the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  solid solution highlighting phonon-glass behaviour as a function of temperature. The theoretical minimum thermal conductivity ( $\kappa_{\min}$ ) of  $\text{SrTiO}_3$  is shown as the dashed line and is calculated from Cahill's model for disordered crystalline solids.<sup>2</sup>



**Fig. S5.** Density of states derived from density functional theory (DFT) level phonon calculations for a)  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  and b)  $\text{SrTiO}_3$ , highlighting the excess density of states in the low frequency region for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ .



**Fig. S6.** Scanning electron micrograph images taken from polished and unpolished surfaces of  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  (a & c, respectively) and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  pellets (b & d, respectively).

**Table S6.** Structural parameters of  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  from Rietveld refinement against SXRD data only (298 K). Space group =  $R\bar{3}c$  with dimensions  $a = b = 5.52589(18)$  Å and  $c = 13.5361(8)$ , and calculated density =  $5.28564(19)$  g cm $^{-3}$ . Refinement statistics are  $R_{\text{wp}} = 3.345$ ,  $R_{\text{exp}} = 1.041$ ,  $\chi^2 = 10.375$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å $^2$ ]
La	6a	0	0	$\frac{1}{4}$	0.5494(6)	0.406(3)
K	6a	0	0	$\frac{1}{4}$	0.4506(6)	0.406(3)
Ti	6b	0	0	0	1.0	0.361(4)
O	18e	0.5109(4)	0	$\frac{1}{4}$	1.0	1.483(13)

**Table S7.** Structural parameters of  $\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). La/K and O sites are modelled using anisotropic displacement parameters. Space group =  $R\bar{3}c$  with dimensions  $a = b = 5.52446(2)$  Å and  $c = 13.53672(3)$  Å, and calculated density =  $5.33471(13)$  g cm $^{-3}$ . Refinement statistics are  $R_{\text{wp}} = 3.901$ ,  $R_{\text{exp}} = 1.779$ ,  $\chi^2 = 15.388$ .

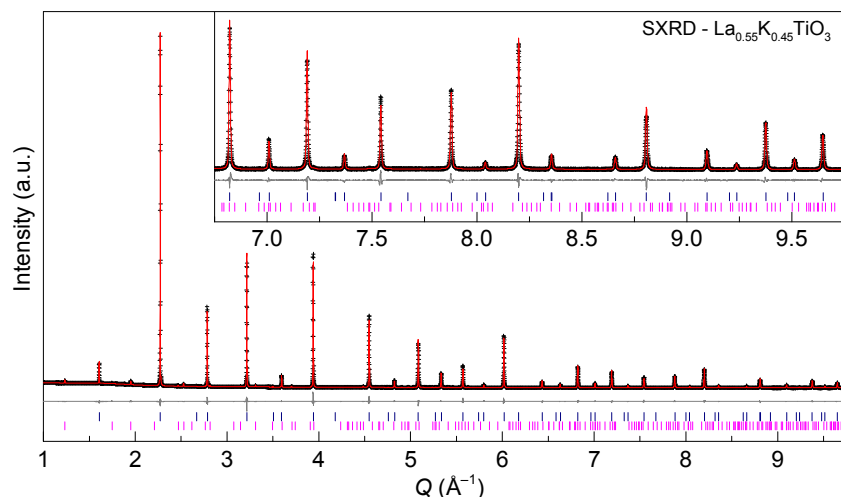
Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å $^2$ ]
La	6a	0	0	$\frac{1}{4}$	0.5786(8)	1.088(9)
K	6a	0	0	$\frac{1}{4}$	0.4214(8)	1.088(9)
Ti	6b	0	0	0	1.00	0.745(16)
O	18e	0.51773(6)	0	$\frac{1}{4}$	1.0005(12)	1.77(9)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
La/K*	0.0045(6)	0.0045(6)	0.0237(16)	0.0022(3)	0	0
O**	0.0159(4)	0.0076(11)	0.0297(12)	0.0038(6)	-0.0048(3)	-0.0097(5)

\* Symmetry conditions for La/K site:  $U_{11}=U_{22}=2U_{12}$ ,  $U_{33}$ , and  $U_{13}=U_{23}=0$

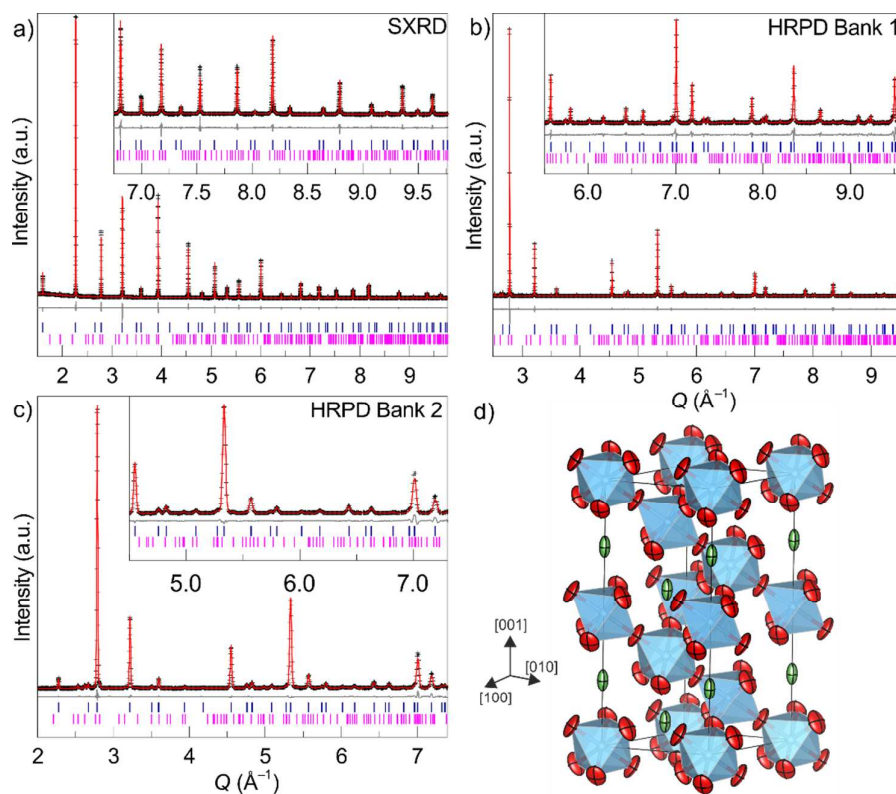
\*\* Symmetry conditions for O site:  $U_{11}$ ,  $U_{22}=2U_{12}$ ,  $U_{33}$ , and  $2U_{13}=U_{23}$

**Table S8.** Bond distances and angles in the  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  series obtained from Rietveld refinement against room temperature SXRD and high resolution NPD data.

y	A–O [Å]	Average A–O [Å]	B–O [Å]	A–O/B–O	B–O–B [°]	$\phi$ [°]
0.45	3 $\times$ 2.7022(17) 6 $\times$ 2.76368(8) 3 $\times$ 2.8237(17)	2.7633(8)	1.95467(11)	1.4137	176.439(10)	2.180(14)
0.4	3 $\times$ 2.6645(3) 6 $\times$ 2.764306(17) 3 $\times$ 2.8604(3)	2.76293(16)	1.95585(2)	1.4127	174.259(19)	3.515(12)



**Fig. S7.** Rietveld refinement against synchrotron X-ray diffraction (SXR) data ( $\lambda = 0.82608$  Å) for  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  using space group  $R\bar{3}c$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up < 5% of the sample.



**Fig. S8.** Rietveld refinements against a) SXR ( $\lambda = 0.82608$  Å) and high resolution neutron powder diffraction (NPD) data measured on b) Bank 1 (backscattering  $2\theta = 168^\circ$ ) and c) Bank 2 ( $2\theta = 90^\circ$  bank) of HRPD at room temperature for  $\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$  using space group  $R\bar{3}c$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up < 2.5% of the sample. The unit cell of  $\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$  is shown in d) with A sites coloured green, B sites coloured blue and oxygen positions in red.

**Table S9.** Structural parameters of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  from Rietveld refinement against SXRD data only (298 K). Space group =  $Pm\bar{3}m$  with dimensions  $a = 3.912089(3)$  Å, and calculated density =  $5.1850(9)$  g cm<sup>-3</sup>. Refinement statistics are  $R_{\text{wp}} = 3.426$ ,  $R_{\text{exp}} = 1.013$ ,  $\chi^2 = 11.438$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	1a	0	0	0	0.5018(7)	0.501(3)
K	1a	0	0	0	0.4982(7)	0.501(3)
Ti	1b	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	0.9576(8)	0.440(4)
Nb	1b	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	0.0424(8)	0.440(4)
O	3c	0	$\frac{1}{2}$	$\frac{1}{2}$	1.0	1.315(12)

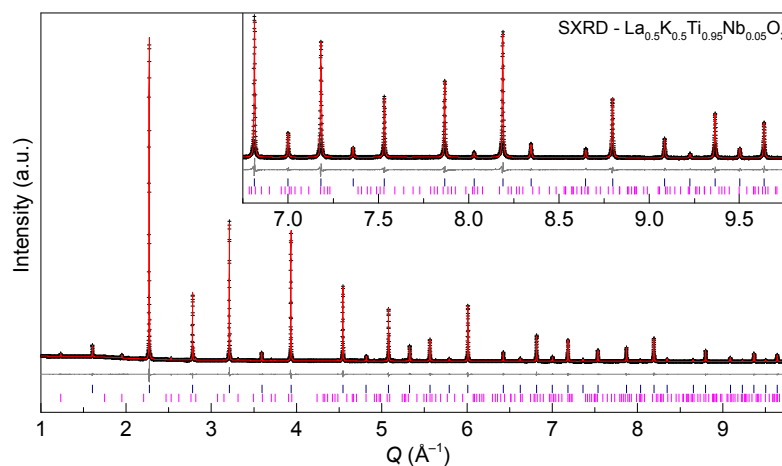
**Table S10.** Structural parameters of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). Coherent neutron scattering lengths: La = 8.24 fm, K = 3.67 fm, Ti = -3.438 fm, Nb = 7.054 fm and O = 5.803 fm.<sup>†</sup> O sites are modelled using anisotropic displacement parameters. Space group =  $Pm\bar{3}m$  with dimensions  $a = 3.915805(12)$  Å, and calculated density =  $5.22537(5)$  g cm<sup>-3</sup>. Refinement statistics are  $R_{\text{wp}} = 1.809$ ,  $R_{\text{exp}} = 0.719$ ,  $\chi^2 = 6.340$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	1a	0	0	0	0.5001(8)	0.777(4)
K	1a	0	0	0	0.4999(8)	0.777(4)
Ti	1b	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	0.9098(6)	0.682(6)
Nb	1b	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	0.0902(6)	0.682(6)
O	3c	0	$\frac{1}{2}$	$\frac{1}{2}$	1.00	1.733(19)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
O	0.0069(3)	0.02252(16)	0.02252(16)	0	0	0

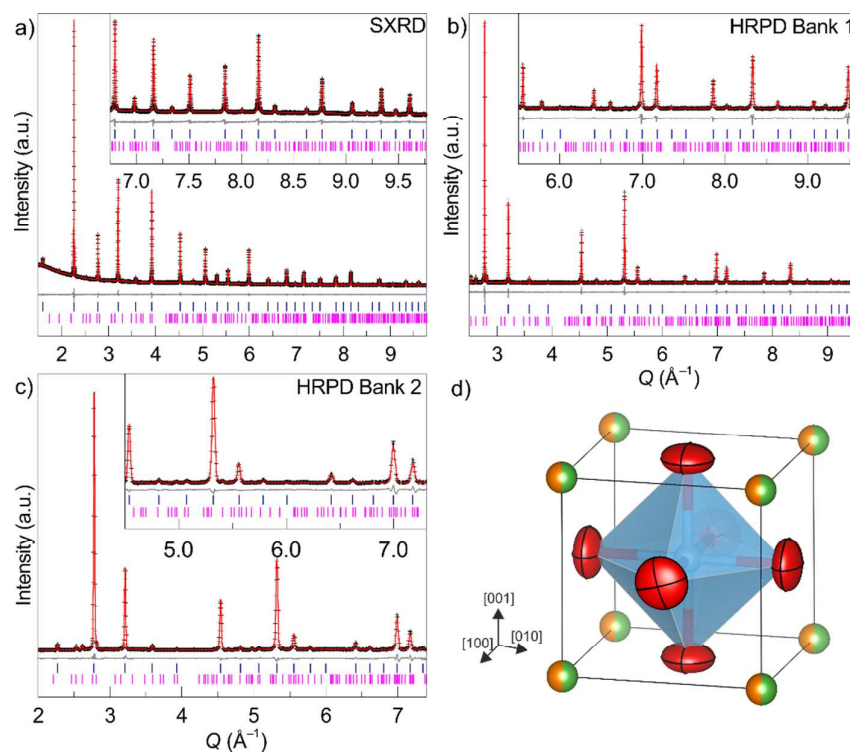
\* Symmetry conditions for O site:  $U_{11}$ ,  $U_{22}=U_{33}$ ,  $U_{12}=U_{13}=U_{23}=0$

**Table S11.** Bond distances and angles in the  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  series obtained from Rietveld refinement against room temperature SXRD and high resolution NPD data.

$z$	A–O [Å]	Average A–O [Å]	B–O [Å]	A–O/B–O	B–O–B [°]	$\phi$ [°]
0.05	$12 \times 2.76626$	2.76626	1.95604	1.4142	180	0
0.10	$12 \times 2.76889$	2.76889	1.95790	1.4142	180	0



**Fig. S9.** Rietveld refinement against synchrotron X-ray diffraction (SXRD) data ( $\lambda = 0.82608$  Å) for  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  using space group  $Pm\bar{3}m$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up  $< 5\%$  of the sample.

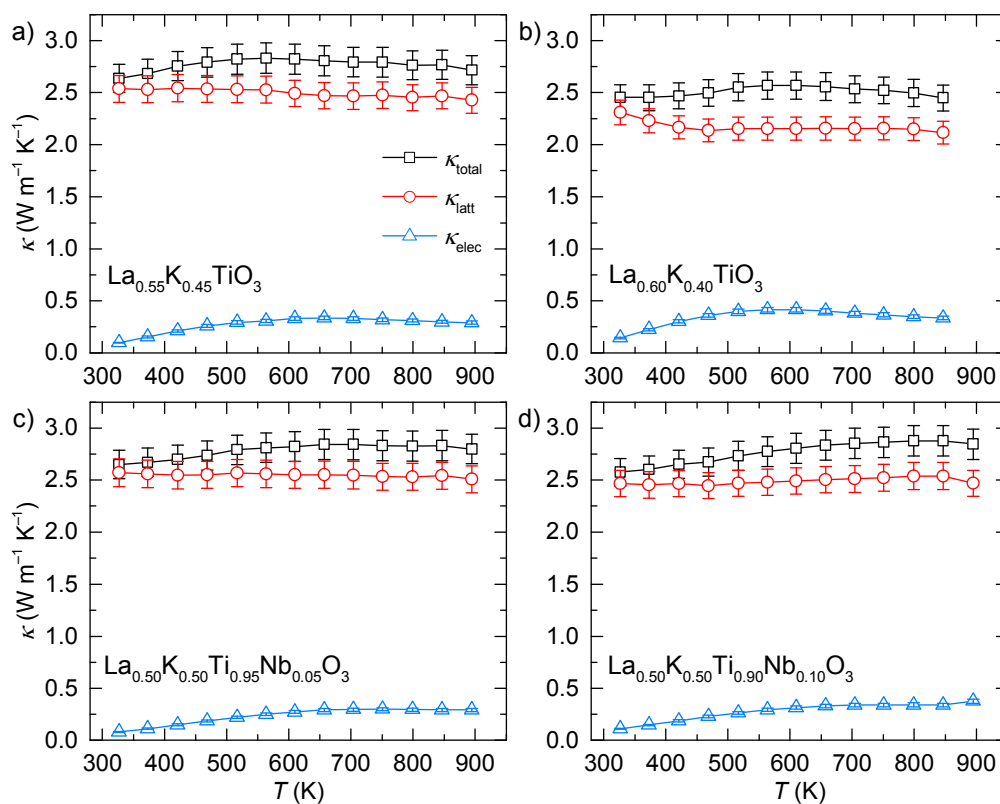


**Fig. S10.** Rietveld refinements against a) SXRD ( $\lambda = 0.82608$  Å) and high resolution neutron powder diffraction (NPD) data measured on b) Bank 1 (backscattering  $2\theta = 168^\circ$ ) and c) Bank 2 ( $2\theta = 90^\circ$  bank) of HRPD at room temperature for  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  using space group  $Pm\bar{3}m$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up  $< 3\%$  of the sample. The unit cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  is shown in d) with A sites coloured green/yellow, B sites coloured blue and oxygen positions in red. Atom sizes are set by  $U_{\text{iso}}$ .

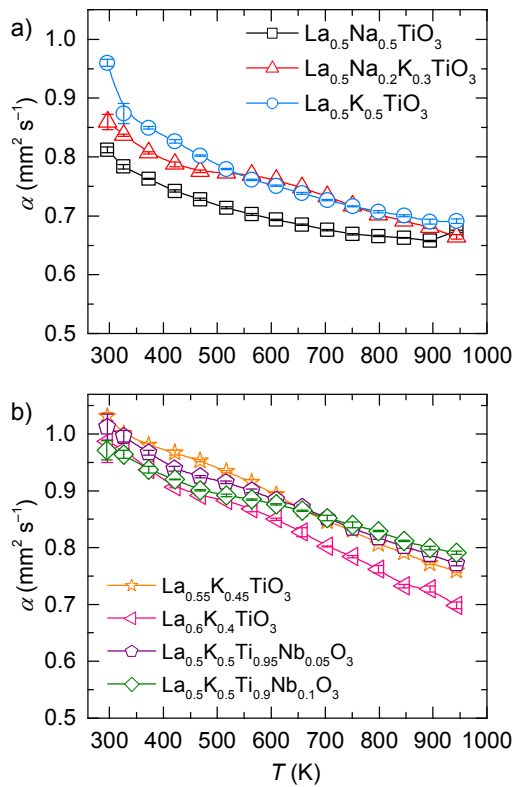


**Table S12.** Theoretical densities ( $\rho_{\text{theo}}$ ) obtained from structural analysis compared with experimental densities ( $\rho_{\text{exp}}$ ) of sintered pellets for each  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  material.

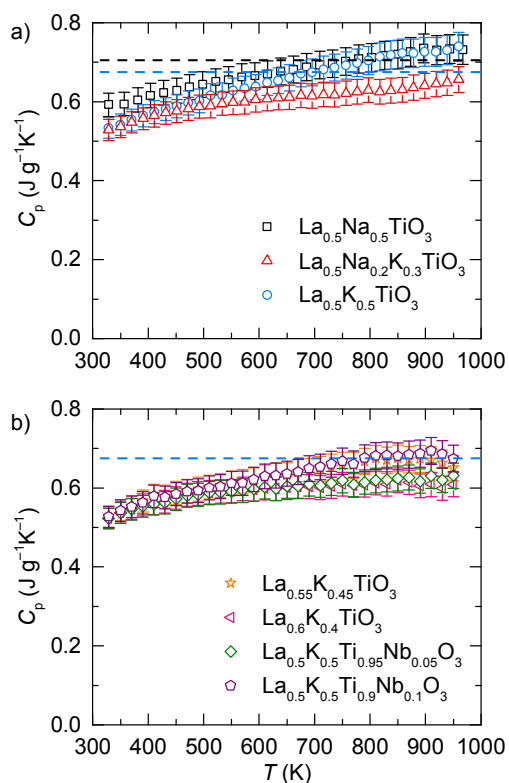
Material	$\rho_{\text{theo}}$ [ $\text{g cm}^{-3}$ ]	$\rho_{\text{exp}}$ [%]
$\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$	5.09091(2)	91.4
$\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$	5.27030(2)	93.9
$\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$	5.33471(13)	90.9
$\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$	5.1850(9)	96.4
$\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$	5.22537(5)	95.9



**Fig. S11.** Temperature dependence of  $\kappa_{\text{total}}$  (black squares),  $\kappa_{\text{latt}}$  (red circles), and  $\kappa_{\text{elec}}$  for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  a)  $y = 0.45$ , b)  $y = 0.4$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  c)  $z = 0.05$ , and d)  $z = 0.1$ . The electronic contributions to the thermal conductivity were calculated through the Wiedemann-Franz law.<sup>4</sup>



**Fig. S12.** Temperature dependence of the thermal diffusivity ( $\alpha$ ) for a)  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  and b) reduced  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions.



**Fig. S13.** Temperature dependence of the specific heat capacity ( $C_p$ ) for a)  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  and b) reduced  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_2\text{O}_3$  compositions. Error bars show 5% uncertainty. Dashed lines represent  $C_p$  values for  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  (black) and  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  (blue) estimated from the Dulong-Petit law,  $C_p = 3nR$ , where  $R$  is the molar gas constant,  $8.314 \text{ J K}^{-1} \text{ mol}^{-1}$  and  $n$  is the number of atoms per unit volume.

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## Journal Name

## ARTICLE

# A and B site doping of a phonon-glass perovskite oxide thermoelectric

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By tuning the A site cation size it is possible to control the degree of octahedral distortion and ultimately structural symmetry in the new perovskite solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ , affording a rhombohedral-to-cubic transition as  $x$  increases above 0.4. The  $\text{La}^{3+}$  and  $\text{K}^+$  cations are distributed randomly across the A site leading to significant phonon disorder in cubic  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  ( $Pm\bar{3}m$ ) which produces a phonon-glass with a thermal conductivity of  $2.37(12) \text{ W m}^{-1} \text{ K}^{-1}$  at 300 K; a reduction of 75% when compared with isostructural  $\text{SrTiO}_3$ . This simple cation substitution of  $\text{Sr}^{2+}$  for  $\text{La}^{3+}$  and  $\text{K}^+$  maintains the flexible structural chemistry of the perovskite structure and two mechanisms of doping for the introduction of electronic charge carriers are explored; A site doping in  $\text{La}_{1-x}\text{K}_x\text{TiO}_3$  or B site doping in  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$ . The phonon-glass thermal conductivity of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is retained upon doping through both of these mechanisms highlighting how the usually strongly coupled thermal and electronic transport can be minimised by mass disorder in perovskites. Precise control over octahedral distortion in A site doped  $\text{La}_{1-x}\text{K}_x\text{TiO}_3$  which has rhombohedral ( $R\bar{3}c$ ) symmetry affords lower band dispersions and increased carrier effective masses over those achieved in B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  which maintains the cubic ( $Pm\bar{3}m$ ) symmetry of the undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  parent. The higher Seebeck coefficients of A site doped  $\text{La}_{1-x}\text{K}_x\text{TiO}_3$  yield larger power factors and lead to increased thermoelectric figures of merit and improved conversion efficiencies compared with the mechanism for B site doping.

## Introduction

Thermoelectric devices convert heat energy directly into electricity or vice versa and have been utilised heavily as power sources for spacecraft and in thermoelectric heating or cooling technologies.<sup>1, 2</sup> The most important application of these devices is to the recovery of waste heat energy, particularly in the industrial, residential and automotive sectors in order to improve efficiency and sustainability.<sup>3, 4</sup> The performance of thermoelectric materials is assessed via the thermoelectric figure of merit  $ZT = (S^2\sigma/\kappa)T$  comprised of the electronic conductivity ( $\sigma$ ) and Seebeck coefficient ( $S$ ), which together form the power factor ( $S^2\sigma$ ), and the thermal conductivity ( $\kappa$ ). Optimisation of  $ZT$  is made difficult by the complex coupled relationships that exist between each of the above properties. The phonon-glass electron-crystal (PGEC) approach envisages the decoupling of  $\sigma$  and  $S$ , which are

typically governed by the Boltzmann transport equation, from  $\kappa$  which is governed by phonon transport within the lattice.<sup>5</sup> "Phonon-glass" refers to the lattice contribution to the thermal conductivity ( $\kappa_{\text{latt}}$ ) being minimised such that heat transport in a material is hindered like in a glass, while "electron-crystal" corresponds to the retention of a high  $\sigma$  and  $S$  within the same material leading to charge transport like in a semiconductor. These characteristics are widely reported in intermetallic systems such as clathrates and skutterudites,<sup>6, 7</sup> which rely on the inclusion of rattler guest atoms within large structural cages to suppress  $\kappa_{\text{latt}}$  while electronic transport mediated through the covalent framework remains unaffected. The PGEC concept has been recently extended to oxide thermoelectrics.<sup>8</sup>

Metal oxides offer increased stability at high operating temperatures (automotive exhaust waste heat recovery typical operation temperature is 350-700 K, and industrial furnace waste heat recovery at 700-1100 K)<sup>4, 9</sup> as well as higher natural abundance and lower toxicity of the constituent elements, leading to much lower costs and smaller environmental impacts when compared with the intermetallics used currently.<sup>10, 11</sup> Titanate-based thermoelectrics, such as donor doped  $\text{SrTiO}_3$ , are the current state-of-the-art n-type thermoelectric oxides with a recent report of  $ZT \approx 0.6$  at 1000 K which was achieved through nanoscale modulation of  $\text{SrTiO}_3$

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† Electronic Supplementary Information (ESI) available. See DOI: 10.1039/x0xx00000x

and  $\text{TiO}_2$  inclusions in  $\text{Sr}_{0.9}\text{La}_{0.1}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$ .<sup>12</sup> The unique band structure of  $\text{SrTiO}_3$  with contributions from both heavy and light carriers at the Fermi level,<sup>13</sup> gives rise to high power factors of  $28\text{--}36\text{ mW K}^{-2}\text{ cm}^{-1}$  at 300 K for  $x = 0.015\text{--}0.1$  in  $\text{Sr}_{1-x}\text{La}_x\text{TiO}_{3-\delta}$  single crystals but  $ZT$  is limited to 0.09 due to the large  $\kappa$  of  $9\text{--}12\text{ W m}^{-1}\text{ K}^{-1}$  in these materials.<sup>14, 15</sup> With  $\kappa_{\text{latt}}$  contributing 75–100% of the total thermal conductivity, control over lattice phonon transport would allow for much higher figures of merit to be achieved in titanate perovskite thermoelectrics.  $\text{SrTiO}_3$  can be considered as a phonon crystal where heat transport is mediated by coherent and long-distance propagation of phonons through the lattice, but due to the relatively small phonon mean free path (2–3 nm) the effects of nanostructuring approaches that are widely used to reduce  $\kappa$  in intermetallic thermoelectrics are negated unless nanometre-sized grains are achieved.<sup>16</sup> We highlighted previously the effectiveness of targeting a low  $\kappa$  that is intrinsic to the material, as opposed to originating from extrinsic microstructural effects, by exploiting the flexibility of the perovskite structure in a simple A site cation substitution that lead to a phonon-glass  $\kappa$  of  $\approx 2.2\text{ W m}^{-1}\text{ K}^{-1}$  in crystalline  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ .<sup>8</sup> Through a combined experimental and computational approach, the origin of this glass-like behaviour was explained by increased phonon scattering as a result of charge and mass disorder resulting from the random distribution of  $\text{La}^{3+}$  and  $\text{Na}^+$ . Specifically, compared against  $\text{SrTiO}_3$  the disordered cations in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  act as locally decoupled oscillators with a broad range of vibrational frequencies which ultimately lead to phonon interactions being far more localised and heat transport being much less coherent. By achieving this intrinsically low  $\kappa$ , it enabled the emergence of PGEC properties where thermal and electronic transport are decoupled and could be altered independently.

The unique band dispersions observed in  $\text{SrTiO}_3$  that allow for high power factors are correlated significantly with the high symmetry of the cubic structure and linearity of the Ti–O–Ti bonding.<sup>13</sup>  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  is a rhombohedrally-distorted perovskite with an  $\approx 8^\circ$  tilt of the  $\text{TiO}_6$  octahedra and non-linear Ti–O–Ti bond angles of  $\approx 167^\circ$ .<sup>8, 17</sup> This tilting decreases overlap interactions between the Ti 3d and O 2p orbitals that are responsible for the dispersion of the valence and conduction bands in  $\text{SrTiO}_3$ , which decreases band dispersion thus increasing the charge carrier effective mass and in turn reducing the electronic conductivity. Symmetry considerations indicate that band dispersions are highest in the cubic perovskite phase, leading to the highest conductivity.<sup>8, 13</sup> It is a widely accepted feature of the perovskite structure that it is possible to influence the symmetry of the material by tuning the cation size, and the effect of this can be estimated through the Goldschmidt tolerance factor,  $t = (r_A + r_O)/(r_B + r_O)$ , where the highest symmetry cubic structures have values closest to unity.<sup>18, 19</sup> By increasing the cation size on the A site of  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  through substitution of  $\text{Na}^+$  for  $\text{K}^+$  ( $r_{\text{Na}^+} = 1.39\text{ \AA}$  and  $r_{\text{K}^+} = 1.64\text{ \AA}$ )<sup>20</sup> in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ ,  $t$  can be increased from 0.98 closer to 1. This is a new solid solution that is not reported in the literature and very little information is available on the fully substituted end-member phase

$\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . This phase has been prepared previously through hydrothermal,<sup>21</sup> and high-pressure synthesis,<sup>22</sup> however detailed analysis and understanding of the structures is lacking. A cubic structure was reported by Brous *et al.*,<sup>23</sup> however this was based on a non-stoichiometric composition of  $\text{La}_{0.531}\text{K}_{0.469}\text{TiO}_3$ . We report the ceramic ambient pressure synthesis and detailed study of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  and the new perovskite solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . Through crystal structure engineering and tuning of A site cation size it is possible to obtain cubic symmetry leading to band dispersions optimised for maximum conductivity in the perovskite  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  while retaining the A site disorder that suppresses  $\kappa$  and leads to phonon-glass behaviour. The chemical flexibility of this material is demonstrated by donor doping on either the A or B site giving rise to electronic transport and power factors comparable to other doped titanate perovskites while the thermal conductivity remains mostly unaffected.

## Experimental

### Synthesis and Processing of $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$

Pre-dried reagents of  $\text{La}_2\text{O}_3$  (99.99%, Aldrich),  $\text{Na}_2\text{CO}_3$  (99.997%, Alfa Aesar),  $\text{K}_2\text{CO}_3$  (99.997%, Alfa Aesar) and  $\text{TiO}_2$  (99.99%, Aldrich) were weighed out on a  $\approx 1.7$  gram scale to give desired compositions. All reagents were dried overnight at 473 K before weighing, except for  $\text{La}_2\text{O}_3$  which was annealed at 1223 K. The powder mixtures were ground in acetone for 10 minutes before being fired as a powder in air at 1273 K for 4 hours placed in an alumina crucible. Heating and cooling rates were 1 and 5  $\text{K min}^{-1}$ , respectively. The fired powders were ground for 10 minutes before being pressed uniaxially into a 13 mm pellet approximately 1 gram in mass. A portion of the powder was kept separate ( $\approx 0.7$  grams) to be used as sacrificial mixture in order to minimise the loss of volatile  $\text{Na}_2\text{O}$  and  $\text{K}_2\text{O}$  species during sintering. The pellet was then further pressed under an isostatic pressure of 200 MPa at ambient temperature using an Autoclave Engineers Cold Isostatic Press. These cold-pressed pellets were placed in an alumina boat and buried in the sacrificial powder before being sintered for 8 hours at 1623 K in air using heating and cooling rates of 1 and 3  $\text{K min}^{-1}$ , respectively.

### Synthesis and Processing of $\text{La}_{1-y}\text{K}_y\text{TiO}_3$ and $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$

Preparation of powder mixtures for the doped materials was done using a procedure similar to that described above for  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ .  $\text{Nb}_2\text{O}_5$  (99.9985%, Alfa Aesar) was used for the B site doped compositions. The sintering step performed at 1623 K for 8 hours, with the cold-pressed pellet buried under sacrificial powder of the same composition, was done under a  $50\text{ mL min}^{-1}$  flow of  $\text{H}_2/\text{N}_2$  (5/95%) instead of air in order to reduce the sample. To ensure the processing conditions were consistent for each composition, the  $p(\text{O}_2)$  was monitored during each reduction step using a Cambridge Sensotec Rapidox 2100 Oxygen Analyser. The furnace was purged with flowing  $\text{H}_2/\text{N}_2$  (5/95%) to achieve a  $p(\text{O}_2)$  of  $10^{-19}$  ppm before the pellets were sintered.

Pairs of pellets for each composition were processed in the same reactions; one pellet for the measurement of thermal transport, and the other for measurement of electronic transport. Pellet densities were determined using the Archimedes method.

### Characterisation

Structural analyses of the new  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ ,  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  materials were performed through Rietveld analysis of synchrotron X-ray powder diffraction (SXRD) data and powder neutron diffraction (PND) data. SXRD data were recorded at beamline I11 ( $\lambda = 0.82608 \text{ \AA}$ ) at Diamond Light Source, U.K, with the powders contained within 0.3 mm diameter borosilicate capillaries. PND data were collected on HRPD at ISIS, the U.K. time-of-flight spallation neutron source, with the powders loaded into thin-walled cylindrical vanadium cans of 8 mm diameter. Data were measured at ambient conditions on detector banks 1 ( $2\theta = 168^\circ$ ), 2 ( $2\theta = 90^\circ$ ) and 3 ( $2\theta = 30^\circ$ ) and were corrected for absorption effects before analysis. Rietveld refinements were carried out using Topas Academic.<sup>24</sup> SEM images were collected using a Hitachi S-4800, cold cathode, field-emission scanning electron microscope (SEM) from pellets that were deposited on a carbon tape attached to an aluminium stub.

### Elemental analysis

Analysis of sample compositions was undertaken using Inductively Coupled Plasma Optical Emission Spectrometry (ICP-OES). Solutions of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  ( $y = 0.5, 0.45$  and  $0.4$ ) and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  ( $z = 0.05$  and  $0.1$ ) were prepared through acid digestion of approximately 10 mg pieces of sintered pellet. These were dissolved in 7 mL of concentrated HCl sealed inside Teflon-lined stainless-steel Parr autoclaves which were heated to 503 K for 4 hours. These solutions were made up to 100 mL in volumetric glassware using ultra-pure water. Metal contents were corrected for interference by comparing separate and combined metal solutions of known concentrations. Data were collected on an Agilent 5110 SCDV ICP-OES instrument.

### Measurement of thermal conductivity

Thermal diffusivity ( $\alpha$ ) data were measured by laser flash analysis using a Netzsch LFA 457. The polished pellets (approximately 10 mm in diameter and 1.5–2 mm thick) were coated in colloidal graphitic carbon before being placed inside the sample chamber which was evacuated and purged three times before being put under a constant helium flow of  $100 \text{ mL min}^{-1}$ . Data were collected in 50 K steps in the temperature range 298 to 943 K, with a five minute equilibration at each temperature. Three measurements were recorded at each temperature and averaged and the standard deviation of these points was typically below 0.5%. Thermal expansion profiles were measured on a Netzsch DIL 402C push-rod dilatometer under dynamic vacuum on bars approximately 4 mm in length. Continuous measurements were made from 298 to 973 K with a heating rate of  $3 \text{ K min}^{-1}$ . Errors on dilatometry measurements are assumed to be 5% as informed by the manufacturer. Heat flux profiles were measured on small pieces of sintered pellet ( $\approx 50 \text{ mg}$ ) using a Netzsch DSC 404 F1

differential scanning calorimeter under a  $50 \text{ mL min}^{-1}$  flow of helium. Data were recorded from 323 to 973 K at a heating rate of  $10 \text{ K min}^{-1}$ . Data were also measured under identical conditions from a sapphire standard (also 50 mg) which was used to determine the heat capacities ( $C_p$ ) of each sample. Errors on heat capacities were assumed to be 5% as advised by the manufacturer. The thermal conductivity ( $\kappa$ ) was calculated by combining the diffusivity, dilatometry and heat capacity data through  $\kappa(T) = \alpha(T) \cdot C_p(T) \cdot \rho(T)$  where  $\rho$  is the pellet density. The total thermal conductivity ( $\kappa$ ) is equal to the sum of the lattice ( $\kappa_{\text{latt}}$ ) and electronic ( $\kappa_{\text{elec}}$ ) contributions through  $\kappa = \kappa_{\text{latt}} + \kappa_{\text{elec}}$ . The electronic contribution to the thermal conductivity is estimated through the Wiedemann-Franz law which states  $\kappa_{\text{elec}} = LT\sigma$ , where  $L$  is the Sommerfeld value for the Lorenz number ( $2.45 \times 10^{-8} \text{ W } \Omega \text{ K}^{-2}$ ),  $T$  is the temperature and  $\sigma$  is the electronic conductivity.<sup>25</sup>

### Measurement of electronic conductivity and Seebeck coefficient

Simultaneous electronic conductivity ( $\sigma$ ) and Seebeck coefficient ( $S$ ) measurements were measured on an Ulvac-Riko ZEM-3 instrument from rectangular bars (dimensions of  $2 \times 2 \times 8 \text{ mm}$ ) which were cut from sintered pellets using a low-speed diamond-blade saw. The bars were mounted in a 4-point geometry with outer current electrodes in contact with each end face and inner thermocouple and voltage probes pressed against one of the longer (8 mm) sides of the bar. The sample chamber was evacuated and purged three times with helium and then dosed with 0.01 MPa of helium before commencing measurement. Data were recorded at 50 K intervals from 323 to 973 K, applying 10, 20 and 30 K temperature gradients to the bar at each temperature step. Errors on both  $\sigma$  and  $S$  were assumed to be 5% on advice of the manufacturer.

### Phonon calculations

To achieve a better understanding of the atomistic origin of the different thermal conductivities of  $\text{La}_{0.5}\text{K}_{0.5}\text{TO}_3$  and  $\text{SrTiO}_3$ , accurate first principles calculations were performed. All phonon calculations of  $\text{La}_{0.5}\text{K}_{0.5}\text{TO}_3$  (with random La/K distribution) and  $\text{SrTiO}_3$  were performed in the harmonic approximation using the code CP2K, at the density functional theory (DFT)/PBE level of theory.<sup>26, 27</sup> CP2K employs a mixed Gaussian/plane-wave basis set; in the current study, a double- $\zeta$  plus polarization quality Gaussian basis set,<sup>28</sup> and a 400 Ry plane-wave cutoff for the auxiliary grid were employed in conjunction with Goedecker–Teter–Hutter pseudopotentials.<sup>29, 30</sup> A convergence threshold of  $1.0 \times 10^{-8} \text{ Ha}$  was used for the self-consistent field cycle, and structural optimizations were considered to have converged when the maximum force on all atoms fell below  $1.0 \times 10^{-4} \text{ Ha Bohr}^{-1}$ . All calculations were performed with the  $\Gamma$ -point approximation, using a rhombohedral supercell containing 640 atoms (lattice constants  $\sim 22 \text{ \AA}$ ). The second derivatives of the total energy with respect to the atomic displacements were computed numerically, from analytical first derivatives, using atomic displacements of 0.01 Bohr from the equilibrium position. Normal modes and frequencies were obtained by diagonalization of the dynamical matrix. To understand the nature of phonon modes in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ , the mean distance

between the 10 atoms that participate to the greatest extent in a given phonon mode as defined by their displacement amplitudes were calculated, denoted as  $d_{\text{average}}$ , which quantifies the localisation of phonon modes and is given by:

$$d_{\text{average}} = \frac{1}{n} \sum_{i,j} d_{ij} \quad (1)$$

where  $i$  and  $j$  are indices of the atoms with the largest displacement amplitudes for the normal mode considered),  $d_{ij}$  is their interatomic distance (using minimum image convention), and  $n$  is the number of pairs of atoms.

#### Electronic band structure calculations

To understand how the octahedral tilting (as present in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  for  $x < 0.4$ ) modifies the band dispersions at the bottom of the conduction band, which may directly affect the thermoelectric power factor, we have performed additional DFT calculations on the electronic band structures of  $\text{SrTiO}_3$ , taking it as a model system, and we consider different amounts of octahedral tilting. All electronic band structure calculations of  $\text{SrTiO}_3$  were performed using CRYSTAL14 at DFT/PBE0 level of theory.<sup>31, 32</sup> Reciprocal space sampling has been performed for all structures using a Monkhorst-Pack grid of  $8 \times 8 \times 8$  k-points. Standard all electron basis sets from the CRYSTAL online database ([www.crystal.unito.it](http://www.crystal.unito.it)) have been used for O and Ti (indicated by the following labels online: O\_8-411d11G\_valenzano\_2006, Ti\_86-411(d31)G\_darco\_unpub). For Sr, we have used a small-core relativistic effective core potential with a double-zeta quality valence basis set (indicated by Sr\_m-S-RSC\_Heyd\_2005\_Heyd\_2005 online). The initial structure of cubic  $\text{SrTiO}_3$  in the rhombohedral cell (space group  $R\bar{3}c$ ) without any octahedral tilting has been fully optimised, which was then used as the starting point to generate  $\text{SrTiO}_3$  structures with octahedral tilting. The two  $\text{SrTiO}_3$  structures with different amounts of octahedral tilting were manually generated, by adjusting the oxygen fractional coordinates and lattice parameters, to ensure Ti–O bond lengths remain the same. Subsequent band structure calculations were performed for the unoptimized structures (for  $\text{SrTiO}_3$  with octahedral tilting).

## Results and Discussion

### Phonon-glass thermal conductivity in $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$

$\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  has a rhombohedral ( $R\bar{3}c$ ) perovskite structure, described by a tilting of the  $\text{TiO}_6$  octahedra about the [111] direction of the primitive cubic cell ( $a^-a^-a^-$  tilt system in Glazer's notation),<sup>33</sup> and disordered A site cation distribution. Phonon calculations that previously highlighted the localisation of vibrational modes in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  were performed for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Calculation of  $d_{\text{average}}$ , which corresponds to the average separation between the ten atoms with largest phonon displacement amplitudes for each individual phonon mode is shown in Fig. 1 for  $\text{SrTiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The results show that  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  has similarly promising features to those observed in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ ,<sup>8</sup>

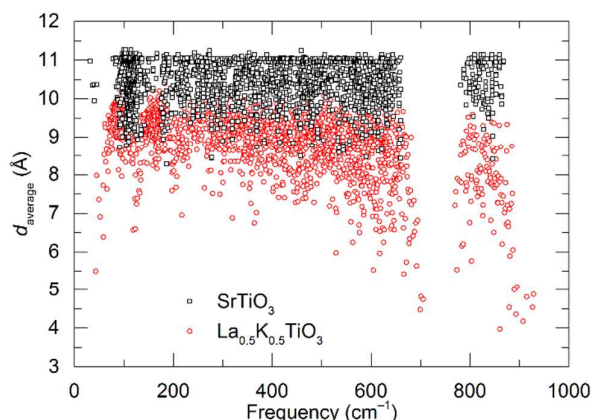


Fig. 1 Phonon localisation in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  (red circles) determined from DFT calculations compared against interactions in the phonon-crystal  $\text{SrTiO}_3$  (black squares). The parameter  $d_{\text{average}}$  is calculated for individual phonon modes as the mean distance between the ten atoms that participate to the greatest extent in that phonon mode as defined by their displacement amplitudes.

indicating that the disordered distribution of  $\text{La}^{3+}$  and  $\text{K}^+$  leads to significant phonon localisation and should yield a phonon-glass thermal conductivity similar to  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ . Compared against  $\text{SrTiO}_3$  which has a  $d_{\text{average}}$  of  $\approx 11$  Å across the entire phonon spectrum,  $d_{\text{average}}$  is reduced by 15–20% in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . We also showed previously how the coexisting dispersive and dispersion-less bands at the bottom of the conduction band that allow for high power factors in cubic  $\text{SrTiO}_3$  are retained in rhombohedral  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ . Table 1 shows relative effective masses ( $m^*$ ) extracted from the calculated electronic structures of  $\text{SrTiO}_3$  at different octahedral tilt angles. The effective masses decrease with increasing tilt angle for both bands, which shows that a cubic zero-tilt structure provides band dispersions that are optimised for highly mobile carriers towards the best possible electronic conductivity. From these computational indications,  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  presents a suitable low thermal conductivity target with band dispersions optimised for electronic conductivity and the structural flexibility of a perovskite to allow for carrier doping.

Potassium was substituted for sodium through the new solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . The results of Rietveld refinement against SXRD and PND data measured on  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ , prepared at 1300 °C through solid state synthesis, are shown in Fig. 2. The single phase powder diffraction pattern for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  can be indexed to a cubic unit cell with space group symmetry  $Pm\bar{3}m$  and lattice parameter of 3.908212(5) Å. This is slightly larger than the only reported lattice parameter of 3.907 Å,<sup>23</sup> which was based on a composition of  $\text{La}_{0.531}\text{K}_{0.469}\text{TiO}_3$  implying that more of the larger  $\text{K}^+$  is present in the structure reported here as evidenced by the refined occupancies (Table S1, ESI) and presents the first structural analysis of stoichiometric  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The  $\text{La}^{3+}$  and  $\text{K}^+$  are modelled on the  $1a$  crystallographic position and both A and B sites are modelled using isotropic displacement

Tilt angle (°)	$m^*_{\text{flat}}$	$m^*_{\text{dispersive}}$	$m^*_{\text{flat}}/m^*_{\text{dispersive}}$
0.00	6.14	1.00/1.00	6.14/6.14
3.25	8.07	1.02/0.99	7.89/8.13
6.50	14.05	1.14/1.05	12.37/13.35



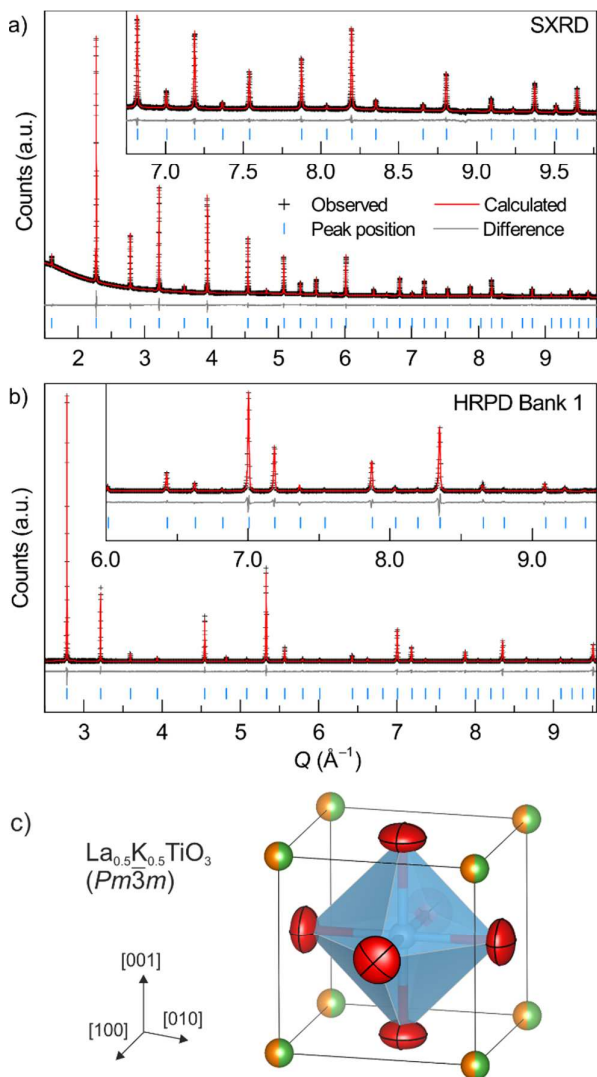


Fig. 2 Rietveld refinements against a) SXRD ( $\lambda = 0.82608 \text{ \AA}$ ) and b) high resolution NPD data measured on Bank 1 (backscattering  $2\theta = 168^\circ$ ) of HRPD at room temperature for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The refinement model is based on cubic  $Pm\bar{3}m$  symmetry with A site metals  $\text{La}^{3+}$  and  $\text{K}^+$  modelled on the  $1a$  (0, 0, 0) position, the  $\text{Ti}^{4+}$  on the  $1b$  ( $\frac{1}{2}, \frac{1}{2}, \frac{1}{2}$ ) position and oxide ions on the  $3c$  positions ( $0, \frac{1}{2}, \frac{1}{2}$ ). The cubic cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is shown in c) with anisotropic displacement ellipsoids for the oxide positions.

parameters (Table S1, Fig. S1, ESI). The use of anisotropic displacement parameters (ADPs) to model the  $3c$  oxygen position yielded a small improvement to

Table 1 Effective masses ( $m^*$ ) for the flat and two dispersive bands along Gamma-Kappa direction extracted from hybrid DFT calculations at different tilt angles in  $\text{SrTiO}_3$ . The effective masses are normalised to the effective mass of the dispersive bands in the cubic structure without octahedral tilting.

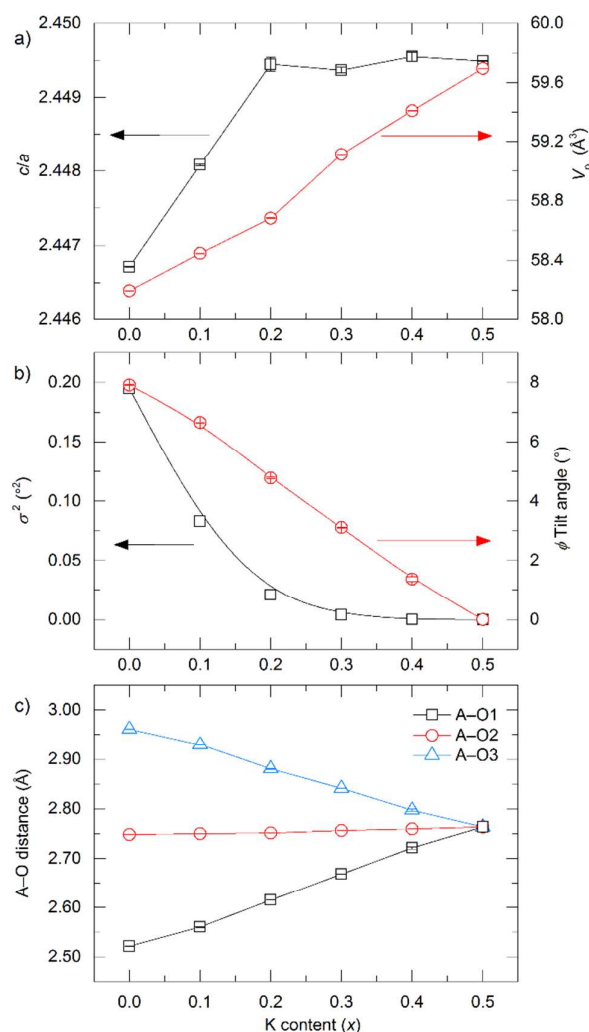


Fig. 3 Variation in a)  $c/a$  lattice parameter ratio (black squares) and primitive pseudocubic cell volume (red circles), b) bond angle variance (black squares) and octahedral tilt angle (red circles), and c) A–O distances as a function of  $x$  in the solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  refined from SXRD and high resolution NPD data. Primitive cell parameters are obtained through  $a_p = a/\sqrt{2}$  and  $c_p = c/2\sqrt{3}$ . Octahedral tilt angles ( $\phi$ ) are determined through  $\phi = \arctan[2\sqrt{3}(x-0.5)]$ , where  $x$  is the  $x$  coordinate of the  $18e$  oxide position in space group  $R\bar{3}c$ . Bond angle variance is obtained from eqn (2) in the text.

the fit, decreasing  $R_{wp}$  from 1.525% to 1.485%. This gave a similar equivalent  $U_{iso}$  ( $0.01446(16) \text{ \AA}^2$ ) to modelling the site using isotropic displacement parameters ( $0.01425(7) \text{ \AA}^2$ ) but resulted in slightly elongated ADPs displaying atomic motion perpendicular to the Ti–O–Ti direction as is expected due to thermally induced rotational motion of the  $\text{BO}_6$  octahedra.

A rhombohedral-to-cubic transition exists within the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  series and occurs upon increasing  $x$  between 0.4 and 0.5 with  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  being cubic ( $Pm\bar{3}m$ ), and compositions of  $0.1 < x < 0.4$  being indexed to the same rhombohedral unit cell as  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  ( $R\bar{3}c$ ). This rhombohedral-to-cubic transition that occurs between  $x = 0.4$  and 0.5 is not immediately obvious upon examination of the  $c/a$  lattice parameter ratio shown in Fig. 3 a) which suggests that the unit cell becomes metrically cubic at  $x = 0.2$  (where  $c/a = \sqrt{6} = 2.4495$ ), and not at  $x = 0.5$ . Only full structural analysis through Rietveld refinement can confirm that the

rhombohedral structure, defined by the symmetry elements present, persists in the  $0.1 < x < 0.4$  compositions and that cubic symmetry is achieved only for  $x = 0.5$ . The primitive pseudocubic cell volume increases linearly towards  $x = 0.5$  as shown in Fig. 3 a). The replacement of smaller sodium with larger potassium ( $r_{\text{Na}^+} = 1.39 \text{ \AA}$  and  $r_{\text{K}^+} = 1.64 \text{ \AA}$ )<sup>20</sup> increases the average cation size on the A site and causes an expansion of the unit cell. This expansion of the unit cell with increasing  $x$  results in a linear decrease in the octahedral tilt angle,  $\phi$ , which is used to describe the rotation of the  $\text{BO}_6$  octahedra about the threefold axis of the primitive cubic cell and is determined from the convention of Megaw,  $\phi = \arctan[2\sqrt{3}(x - 0.5)]$ , where  $x$  is the positional parameter of the 18e oxygen site.<sup>34</sup> Fig. 3 b) shows how the octahedral tilt angle decreases from  $7.9142(2)^\circ$  for  $x = 0$  to  $0^\circ$  for  $x = 0.5$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ , with the rhombohedral-to-cubic transition between tilted and non-tilted structures occurring between  $x = 0.4$  and  $0.5$ . Alternatively, this can be viewed through the Ti–O–Ti bond angles which become more linear with  $x$ , increasing from  $167.044(14)^\circ$  for  $x = 0$  to  $180^\circ$  in  $x = 0.5$ . While  $\phi$  is decreasing towards  $0^\circ$  with increasing  $x$ , there is a simultaneous distortion of the octahedra themselves where O–Ti–O angles deviate away from the ideal  $90^\circ$  and can be quantified by the octahedral bond angle variance in eqn (2)<sup>35</sup>

$$\sigma^2 = \frac{1}{n-1} \sum_i^n (\theta_i - \theta_0)^2 \quad (2)$$

where  $n$  is the number of bonds while  $\theta_i$  and  $\theta_0$  denote the  $i^{\text{th}}$  and ideal bond angles of the octahedron, respectively. This is shown to decrease in Fig. 3 b) as a function of increasing  $x$ , reaching zero at  $x = 0.5$  indicating ideal  $90^\circ$  O–Ti–O angles are achieved in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Although it is not fully understood, it is evident that this competition between simultaneous octahedral tilting ( $\phi$ ) and octahedral distortion ( $\sigma^2$ ) in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  leads to the compositions of  $0.2 < x < 0.4$  adopting metrically cubic unit cells. Fig. 3 c) shows that the separate A–O distances converge at  $x = 0.5$  with the average A–O distance increasing following the expansion of the A site cations and unit cell. The A–O/B–O ratio increases with  $x$  towards the value of  $\sqrt{2}$  (Table S4, ESI) which relates these distances in the cubic perovskite structure. The cubic structure of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is very similar to that of  $\text{SrTiO}_3$  with comparable lattice parameters of  $a = 3.908212(5) \text{ \AA}$  and  $3.909149(3) \text{ \AA}$ , respectively.<sup>36</sup> Refinement of ADPs to model the 18e oxygen positions in rhombohedral  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  gave similar results to those in cubic  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  where their positions were elongated normal to the central atom of the octahedron and along the direction of octahedral rotation (Fig. S1, ESI). The symmetry of space group  $R\bar{3}c$  allows for refinement of ADPs for the A site, which results in a slight elongation of the position along the  $c$  axis of the unit cell towards neighbouring octahedra (Fig. S1, ESI). Despite the static disorder generated by three different cations occupying the A site, the local deviation away from the average structure is minimal with a root mean square displacement of  $0.107(16) \text{ \AA}$  at 300 K. Most importantly, the structural analysis of the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$

solid solution and of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  in particular, shows that the  $\text{La}^{3+}$  and  $\text{K}^+$  are disordered across the twelve-coordinate A site. This disorder induced phonon-glass behaviour of the thermal conductivity in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ ,<sup>8</sup> a property expected to be replicated in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  following the observation of phonon localisation (Fig. 1).

$\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is a crystalline material that offers the high symmetry cubic structure of  $\text{SrTiO}_3$ , with the unique multi-carrier band structure suited towards thermoelectricity,<sup>13</sup> along with a disordered distribution of A

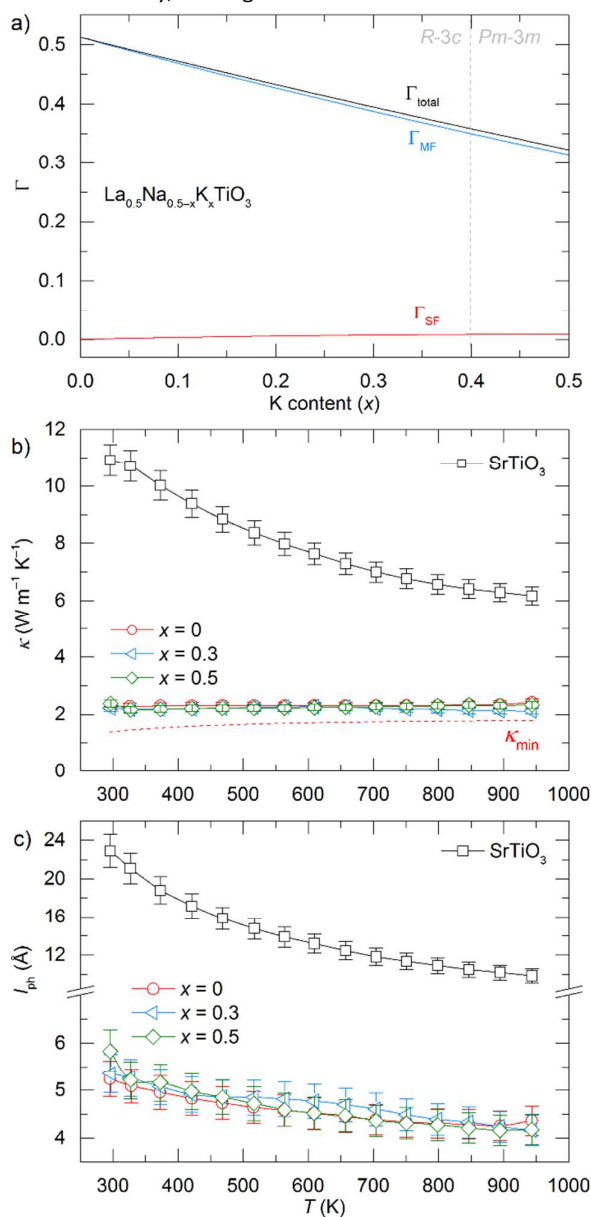


Fig. 4 Dependence of a) disorder scattering parameter ( $l$ ), including mass ( $l_{\text{MF}}$ ) and size ( $l_{\text{SF}}$ ) contributions shown in blue and red respectively, b) thermal conductivity ( $k$ ), and c) phonon mean free path ( $l_{\text{ph}}$ ) on  $x$  in the solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . The theoretical minimum thermal conductivity ( $k_{\text{min}}$ ) of  $\text{SrTiO}_3$  is shown as the dashed line in b) and is calculated from Cahill's model for disordered crystalline solids. [REF-Cahill1992] Phonon mean free paths in c) are obtained through eqns (4) and (5).

site metals which have considerable mass contrast. The disorder scattering parameter is defined by  $\Gamma = \Gamma_{\text{MF}} + \Gamma_{\text{SF}}$  and contributes significantly to phonon scattering in these materials. It combines contributions from the mass contrast of defects ( $\Gamma_{\text{MF}}$ ) and the size strain ( $\Gamma_{\text{SF}}$ ) induced by ionic radius variance of defects, both of which are effective disruptors of phonon transport within a crystal lattice.<sup>37</sup> The mass fluctuation and strain field terms are quantified through  $\Gamma_{\text{MF}} = \sum_i f_i (1 - m_i/\bar{m})^2$  and  $\Gamma_{\text{SF}} = \sum_i f_i (1 - r_i/\bar{r})^2$  respectively, where  $f_i$  is the fraction of atoms which have mass  $m_i$  and radius  $r_i$  occupying a crystallographic site with average mass  $\bar{m}$  and average radius  $\bar{r}$ . Fig. 4 a) shows how  $\Gamma$ ,  $\Gamma_{\text{MF}}$  and  $\Gamma_{\text{SF}}$  change as a function of  $x$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . Replacement of lighter sodium with heavier potassium results in  $\Gamma_{\text{MF}}$  decreasing from 0.513 for  $x = 0$  to 0.314 for  $x = 0.5$ , while the increased size disparity between  $\text{La}^{3+}$  and  $\text{K}^+$  raises  $\Gamma_{\text{SF}}$  from  $1.19 \times 10^{-4}$  to  $8.71 \times 10^{-3}$  for those respective values of  $x$ . Although  $\Gamma_{\text{SF}}$  is no longer negligible, with a 2.70% contribution to  $\Gamma$  in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  compared to 0.02% in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ , this increase in cation size variance is insufficient to alleviate the reduction in  $\Gamma_{\text{MF}}$  that remains the dominant contribution to the total  $\Gamma$ , which decreases from 0.513 to 0.323 between  $x = 0$  and 0.5, respectively. The measured thermal conductivities across the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  series are shown in Fig. 4 b) (Fig. S4, ESI). One of the more striking features of the experimental data is that the thermal conductivity of the compositions in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are not significantly different despite the reduction in  $\Gamma$  shown in Fig. 4 a). The compositions of  $x = 0, 0.3$  and  $0.5$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are electronically insulating meaning that  $\kappa$  is equal to  $\kappa_{\text{latt}}$  in these materials. The thermal conductivity of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  retains the phonon-glass characteristics that were observed for  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  where the temperature dependence of  $\kappa_{\text{latt}}$  matches that of the heat capacity. Additionally, both the magnitude and temperature dependence of  $\kappa_{\text{latt}}$  observed for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  are comparable with the theoretical minimum thermal conductivity ( $\kappa_{\text{min}}$ ) that would be expected if  $\text{SrTiO}_3$  exhibited phonon glass properties. This quantity, compared against experimental data in Fig. 4 b), is estimated through Cahill's model for disordered crystalline solids described by eqn (3)

$$\kappa_{\text{min}} = \left(\frac{\pi}{6}\right)^{\frac{1}{3}} k_B N^{\frac{2}{3}} \sum_i v_i \left(\frac{T}{\theta_i}\right)^2 \int_0^{\frac{\theta_i}{T}} \frac{x^3 e^x}{(e^x - 1)^2} dx \quad (3)$$

where  $k_B$  is the Boltzmann constant,  $N$  is the number density of atoms known from structural refinement,  $v_i$  is the velocity of sound and  $\theta_i$  is used as the cut-off frequency.<sup>38</sup> Experimental values of transverse and longitudinal velocities of sound extracted from  $\text{SrTiO}_3$  single crystals were used to calculate  $\theta_i$  through eqn (4) and are listed in Table S5 (ESI)<sup>39</sup>

$$v_s = \frac{k_B \theta_D}{h \sqrt[3]{6\pi^2 N}} \quad (4)$$

The close comparison to  $\kappa_{\text{min}}$  and glass-like dependence of  $\kappa_{\text{latt}}$  with temperature shows that  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is a crystalline material with phonon glass thermal properties. The magnitude of  $\kappa_{\text{latt}}$  at room temperature increases from 2.28 to 2.40  $\text{W m}^{-1}$

$\text{K}^{-1}$  from  $x = 0$  to 0.5 respectively, which is associated with the decrease in  $\Gamma$ . This is only a minor change in  $\kappa_{\text{latt}}$  in contrast to a considerable change to  $\Gamma$ , indicating that the disorder parameter has a smaller influence on the thermal conductivity in materials which display phonon-glass properties compared to those with phonon-crystal properties as was shown in  $\text{Sr}_{1-x}\text{La}_{0.67x}\text{TiO}_3$  where increasing  $\Gamma$  had a smaller effect as  $\kappa_{\text{latt}}$  approached  $\kappa_{\text{min}}$ .<sup>40</sup> The values of  $\kappa_{\text{latt}}$  observed for  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are very low when compared against both doped and undoped compositions of  $\text{SrTiO}_3$  which exhibit  $\kappa_{\text{latt}}$  values in the range of 4 – 12  $\text{W m}^{-1} \text{K}^{-1}$  at 300 K,<sup>41, 42</sup> and perhaps most importantly exhibit a temperature dependence that is typical of a phonon-crystal where a  $T^{-1}$  relationship is observed at high temperatures due to the anharmonic interaction of phonon-phonon Umklapp scattering processes. Data measured from undoped  $\text{SrTiO}_3$  is provided for comparison purposes in Fig. 4 b) and show that the substitution of both  $\text{La}^{3+}$  and  $\text{K}^+$  for  $\text{Sr}^{2+}$  on the A site of the perovskite structure affords a 75% reduction in thermal conductivity at 300 K (2.37(12)  $\text{W m}^{-1} \text{K}^{-1}$  for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  compared against 10.91(56)  $\text{W m}^{-1} \text{K}^{-1}$  for  $\text{SrTiO}_3$ ). This low, glass-like thermal conductivity observed experimentally for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is consistent with the results obtained from calculations (Fig. 1) which showed that phonons are localised in comparison with the phonon-crystal  $\text{SrTiO}_3$ . The calculations, performed at DFT level, show increased contributions to the phonon density of states (PDOS) in the low frequency region in comparison with  $\text{SrTiO}_3$  (Fig. S5, ESI). The largest contribution to the PDOS in  $\text{SrTiO}_3$  is at  $\approx 110 \text{ cm}^{-1}$ , however two broader contributions are observed in this region for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ ; a peak at  $\approx 95 \text{ cm}^{-1}$  predominantly consisting of contributions from La, and a second peak at  $\approx 170 \text{ cm}^{-1}$  arising from the contributions of K and Ti. Through this combined experimental and computational understanding of phonon behaviour in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ , it is clear the vibrational disorder introduced by having two high-mass contrast cations present on a single crystallographic position is highly effective at scattering phonons and reducing heat transport within the lattice, leading to the observed glass-like thermal conductivity in crystalline  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . This increased localisation of phonons is similar to that observed in  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ ,<sup>8</sup> and shows that the disordered  $\text{La}^{3+}$  and  $\text{K}^+$  cations in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  are responsible for the phonon-glass behaviour by acting as decoupled local oscillators with incoherent phases which dramatically reduce time and distance between phonon scattering events. Experimentally, this is evident from the phonon mean free paths that are shown in Fig. 4 c) and are calculated from eqns (4) and (5) using  $\theta_D$  of 630 K.<sup>43</sup>

$$\kappa_{\text{latt}} = \frac{1}{3} \cdot C_V \cdot l_{\text{ph}} \cdot v_s \quad (5)$$

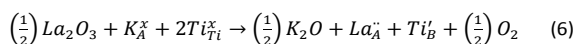
where  $C_V$  is the isochoric heat capacity, and  $l_{\text{ph}}$  is the phonon mean free path. The phonon mean free paths of compositions in the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  series all show relatively flat temperature dependence in comparison with  $\text{SrTiO}_3$  which is a clear indicator of the emergence of glassy characteristics in these materials. The magnitudes of  $l_{\text{ph}}$  in  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  of 4-

6 Å are comparable to those of vitreous materials such as SiO<sub>2</sub> which typically are around 5–7 Å.<sup>44</sup> Microstructure was shown to influence  $\kappa_{\text{latt}}$  and  $I_{\text{ph}}$  of SrTiO<sub>3</sub>,<sup>16</sup> however, this required nanometre-sized grains. The effect of microstructure in La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> materials is negligible due to the large disparity between  $I_{\text{ph}}$  and grain size, which are shown through SEM (Fig. S6, ESI) to be much larger and of the order of 10–20 µm.

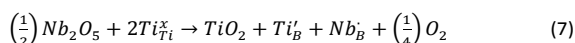
#### Introduction of electronic charge carriers into La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub>

One of the more useful features resulting from the complete replacement of Sr<sup>2+</sup> for a charge balanced mixture of La<sup>3+</sup> and K<sup>+</sup> in La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> is that the chemical and structural flexibility of the perovskite structure is maintained. Electron doping in SrTiO<sub>3</sub> involves the incorporation of an aliovalent cation with higher charge, such as Ln<sup>3+</sup> on the A site or Nb<sup>5+</sup> on the B site,<sup>42, 45, 46</sup> which through charge compensation mechanisms reduces a sufficient amount of Ti<sup>4+</sup> to Ti<sup>3+</sup> giving rise to electronic transport. Even though the A site in La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> has a disordered cation arrangement, the flexible structural chemistry enables the introduction of electronic carriers through doping of either the A or B site. Incorporation of Nb<sup>5+</sup> onto the B site through La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> allows for the mechanisms that govern the thermal and electronic properties to be confined to the separate A and B sites of the perovskite structure, respectively, as was reported for PGEC La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>1–x</sub>Nb<sub>x</sub>O<sub>3</sub>.<sup>8</sup> Doping of the A site can be achieved by targeting non-stoichiometric compositions through La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> (with  $y < 0.5$ ). This maintains the mass contrast that affords the phonon-glass thermal conductivity, and usefully allows for the same nominal charge carrier concentrations as La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> to be accessed by using smaller dopant amounts due to the nature of the dopant mechanisms which are described by Kröger-Vink notation in reaction eqns (6) and (7), where A and B refer to the site occupied.

For La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub>:



For La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub>:



Through solid state reaction performed at 1300 °C under a H<sub>2</sub>/N<sub>2</sub> (5%) atmosphere, it is possible to produce phase pure samples of La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> with  $0.4 \leq y \leq 0.5$  and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> up to  $z = 0.1$ . These values define the limits of doping for each material under these particular synthetic conditions; decreasing  $y$  below 0.4 in La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> results in the formation of La<sub>2</sub>Ti<sub>2</sub>O<sub>7</sub> impurities, while increasing  $z$  further than 0.1 yields LaNbO<sub>4</sub> as an impurity phase in La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub>. The effects of A and B site doping upon the unit cell parameters of La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> are shown in Figs. 5 a) and b), respectively. A linear decrease in unit cell volume is observed as a function of La<sup>3+</sup> content in La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> associated with the larger K<sup>+</sup> cation being replaced with La<sup>3+</sup> ( $r_{\text{K}^+} = 1.64$  Å and  $r_{\text{La}^{3+}} = 1.36$  Å).<sup>20</sup> Both La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> compositions are indexed to rhombohedral  $R\bar{3}c$  symmetry with separation of primitive cell parameters  $a_p$  and  $c_p$ , shown in Fig. 5 a) (Tables S6, S7 and S8,

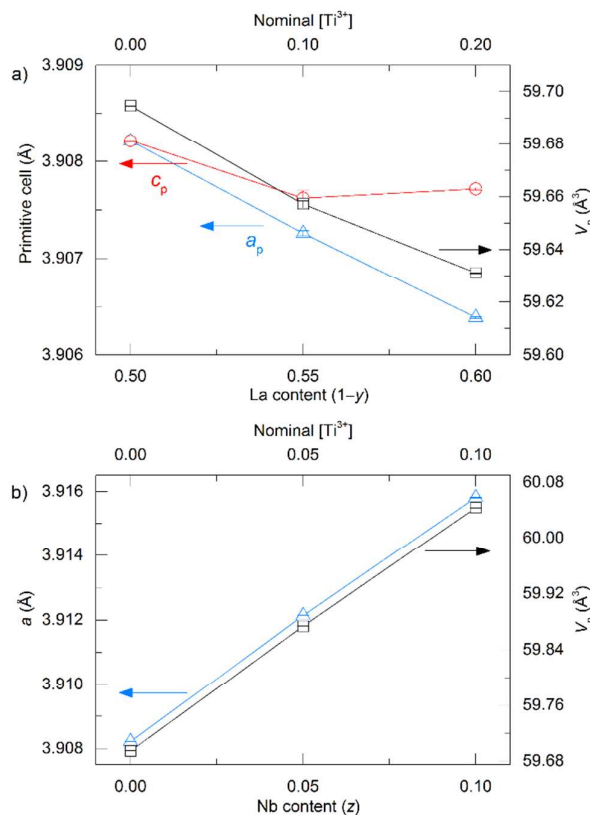


Fig. 5 Variation in a) primitive unit cell parameters and volume as a function of La<sup>3+</sup> content in A site doped La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> with  $R\bar{3}c$  symmetry, and b) cubic unit cell parameter and volume in B site doped La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> ( $Pm\bar{3}m$ ) as a function of Nb<sup>5+</sup> substitution. The nominal Ti<sup>3+</sup> concentrations are plotted on the top axes of each and increase with increasing La<sup>3+</sup> and Nb<sup>5+</sup> content. The La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> series are cubic ( $Pm\bar{3}m$ ), while La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub> materials are rhombohedral ( $R\bar{3}c$ ).

Figs. S7 and S8, ESI). This distortion away from the cubic symmetry of the La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> parent results from the decreasing average cation size on the A site, as well as an increasing cation size on the B site from the reduction of Ti<sup>4+</sup> to Ti<sup>3+</sup> ( $r_{\text{Ti}^{4+}} = 0.605$  Å and  $r_{\text{Ti}^{3+}} = 0.67$  Å); the tolerance factor can be reduced by decreasing  $r_A$  or increasing  $r_B$ . Contrary to this, the cubic symmetry of the La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub> parent is maintained in the B site doped La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1–z</sub>Nb<sub>z</sub>O<sub>3</sub> materials up to  $z = 0.1$  with a linear increase in the unit cell parameter and volume shown in Fig. 5 b) (Tables S9, S10 and S11, Figs. S9 and S10, ESI), consistent with the substitution of Ti<sup>4+</sup> with larger Nb<sup>5+</sup> ( $r_{\text{Ti}^{4+}} = 0.605$  Å and  $r_{\text{Nb}^{5+}} = 0.64$  Å). There remains no evidence for ordering of the A site cations even in non-stoichiometric La<sub>1–y</sub>K<sub>y</sub>TiO<sub>3</sub>, indicating that the disordered distributions of La<sup>3+</sup> and K<sup>+</sup>, which we have shown to be essential for the phonon glass  $\kappa$  in these materials, are preserved in all materials reported in this study. The refined oxygen occupancies for the A and B doped materials studied through both SXRD and NPD did not provide any evidence for oxide deficiencies (Tables S7 and S10,

Nominal composition	La	K	Ti	Nb
La <sub>0.5</sub> K <sub>0.5</sub> TiO <sub>3</sub>	0.518(3)	0.507(5)	1.000(5)	–
La <sub>0.55</sub> K <sub>0.45</sub> TiO <sub>3</sub>	0.556(3)	0.445(2)	1.000(5)	–
La <sub>0.6</sub> K <sub>0.4</sub> TiO <sub>3</sub>	0.608(3)	0.413(2)	1.000(7)	–
La <sub>0.5</sub> K <sub>0.5</sub> Ti <sub>0.95</sub> Nb <sub>0.05</sub> O <sub>3</sub>	0.505(2)	0.493(4)	0.950(3)	0.0487(1)
La <sub>0.5</sub> K <sub>0.5</sub> Ti <sub>0.9</sub> Nb <sub>0.1</sub> O <sub>3</sub>	0.507(7)	0.469(2)	0.900(6)	0.0981(1)

ESI). When refined, values within error of a fully occupied oxygen position were obtained much like the oxygen occupancies in the undoped materials  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  and  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  (Tables S1 and S2, ESI).

The metal contents for each of the doped materials were determined using ICP-OES and are listed in Table 2. Both the La/K and Ti/Nb ratios agree very well with those from the nominal compositions, and compare closely to the refined occupancies obtained from Rietveld analysis of SXRD and PND data (Tables S1, S6, S7, S9 and S10, ESI). The analysis shows near-stoichiometric ratios for La and K in the undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  parent, in contrast to the single previous report of cubic  $\text{La}_{0.531}\text{K}_{0.469}\text{TiO}_3$ , in which not all of the alkali metal was incorporated into the structure.<sup>23</sup>

Table 2 Results of ICP-OES measurements on  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  materials. Values are normalised to titanium and compared against the nominal expected compositions. Metal contents were corrected by analysing metal solutions of known concentrations both separately and combined. Correction factors are La = 1.01, K = 1.10, Ti = 1.00 and Nb = 1.00. Absolute errors are shown.

The sintering of cold-pressed pellets of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions at 1350 °C under a  $\text{H}_2/\text{N}_2$  (5%) atmosphere yielded ceramics with measured densities higher than 90% of the values expected from structural analysis (Table S12, ESI), and SEM imaging showed the pellets to have minimal porosity (Fig. S6, ESI).

The temperature dependences of the thermal conductivity for doped  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions are shown in Fig. 6 a). These materials exhibit low phonon-glass thermal conductivities comparable to that of undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  despite possessing electron carrier concentrations, determined from nominal compositions, which lie within the range expected for heavily doped semiconductors ( $10^{19}$ - $10^{21} \text{ cm}^{-3}$ ). The thermal conductivities have magnitudes of 2.45(12)-2.65(13)  $\text{W m}^{-1} \text{ K}^{-1}$  at room temperature and 2.45(12)-2.80(15)  $\text{W m}^{-1} \text{ K}^{-1}$  at 850 K and have a temperature dependence like  $\kappa_{\text{min}}$  determined from eqn (3). In contrast to the thermal conductivities of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  which show no correlation with doping, the electronic conductivities and Seebeck coefficients display a strong doping dependence shown in Figs. 6 b) and c). In  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ , peak conductivities increase from 180(9) to 212(10)  $\text{S cm}^{-1}$  at 600 K for  $z = 0.05$  and 0.1 respectively, whereas the peak values of  $\sigma$  in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  increase from

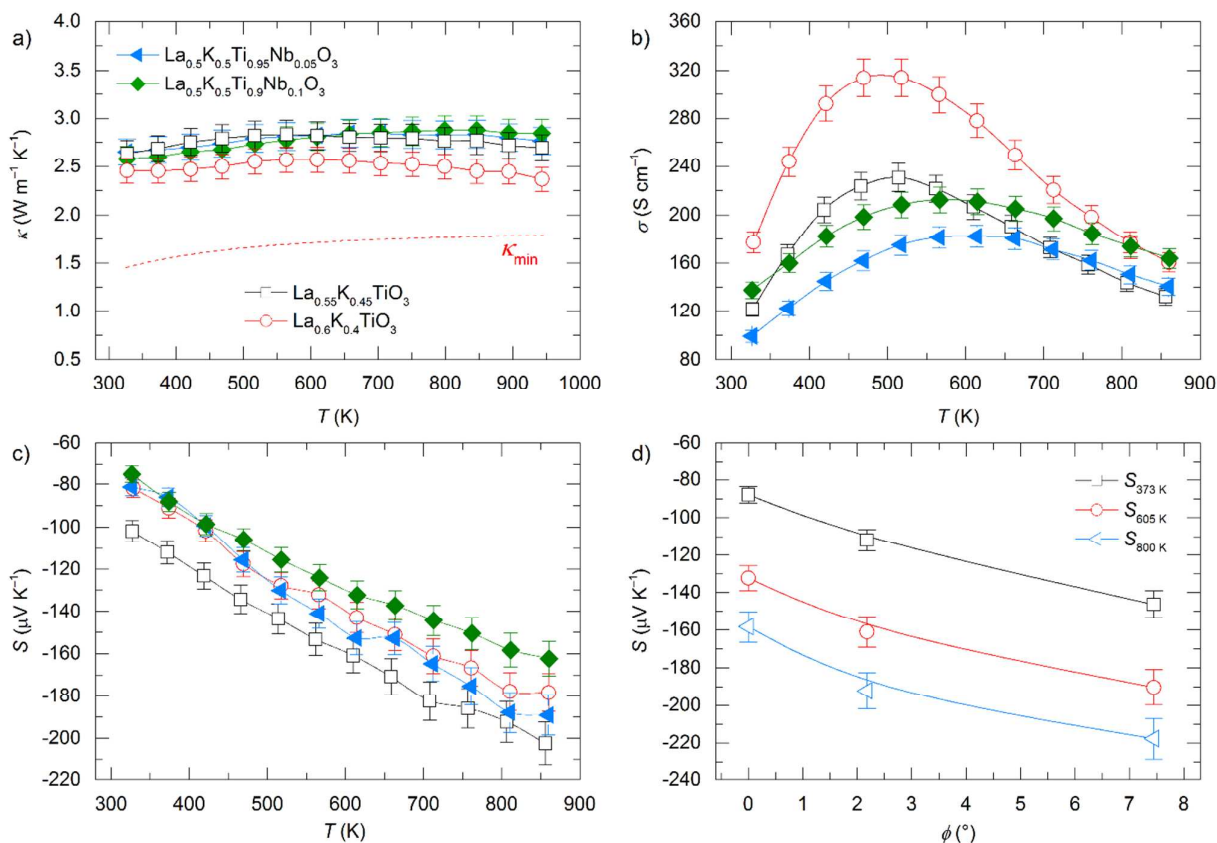


Fig. 6 Temperature dependence of a) thermal conductivity ( $\kappa$ ), b) electronic conductivity ( $\sigma$ ), and c) Seebeck coefficient ( $S$ ) for A site ( $\text{La}_{1-y}\text{K}_y\text{TiO}_3$ ,  $y = 0.45$  and  $0.4$ ) and B site ( $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ ,  $z = 0.05$  and  $0.1$ ) doped compositions of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Dependence of  $S$  as a function of tilt angle ( $\phi$ ) at different temperatures is shown in d) using three materials that have the same nominal carrier concentration:  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  ( $\phi = 0^\circ$ ) and  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  ( $\phi = 2.18(1)^\circ$ ) from this study, and  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  ( $\phi = 7.45(3)^\circ$ ) which was taken from a recent study of PGEC  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  materials with  $R3c$  symmetry.[REF-Daniels2017]



230(11) S cm<sup>-1</sup> for  $y = 0.45$  to 315(16) S cm<sup>-1</sup> for  $y = 0.4$  at 500 K, associated with increasing carrier concentration. The observed temperature dependencies of both  $\sigma$  and  $S$  for all compositions are similar to many doped titanate thermoelectrics reported in the literature.<sup>40, 42, 46, 47</sup> Maxima in  $\sigma$  are observed between 500 – 600 K and are associated with increased resistance from acceptor-based phases which are localised to the grain boundaries.<sup>48</sup> The negative values of  $S$  indicate electron doping of predominantly  $n$ -type carriers within all compositions, and both A and B site doped materials exhibit decreasing  $|S|$  with increasing carrier concentration which is expected for metals or parabolic band semiconductors through eqn (8)

$$S = \frac{8\pi^2 k_B^2}{3eh^2} m^* T \left( \frac{\pi}{3n} \right)^{2/3} \quad (8)$$

where  $h$  is Planck's constant,  $e$  is elementary charge,  $m^*$  is the carrier effective mass and  $n$  is the carrier concentration.<sup>3</sup> For La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub>, which nominally have the same  $n$ , almost identical conductivities of  $\approx 210(10)$  S cm<sup>-1</sup> are observed at 600 K. Interestingly for these particular compositions,  $|S|$  is considerably higher for La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> than for La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub>, with values of  $-161(8)$  and  $-132(7)$   $\mu$ V K<sup>-1</sup> at 600 K respectively, and  $|S|$  of La<sub>0.6</sub>K<sub>0.4</sub>TiO<sub>3</sub> is still higher than that of La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> despite having a higher nominal carrier concentration. As La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> both have the same  $n$  but a larger  $|S|$  is observed experimentally for La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub>, from the relationship shown in eqn (8) this indicates that a higher  $m^*$  is achieved through the A site doping of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> compared to B site doping through La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. This shows how the slight differences in the crystal structures of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub> can significantly influence the electronic properties. In general, 3d-2p orbital overlap in these perovskites is maximised through shorter bonding and/or tilt angles that approach 0°. In La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub> the Nb<sup>5+</sup> substitution retains the linear Ti–O–Ti bonding of the cubic ( $Pm\bar{3}m$ ) structure (Table S11, ESI). This maximises band dispersion and therefore  $\sigma$ , but leads to a lower  $m^*$  which supports the decreased  $|S|$  observed experimentally for La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. In contrast, the change of symmetry to  $R\bar{3}c$  in La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> reduces the orbital overlap through octahedral tilting which reduces band dispersion yielding a larger  $m^*$  and is the reason that higher  $|S|$  values are observed for La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> compared to La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. The decrease in  $\sigma$  that a larger  $m^*$  is expected to bring about is offset by the increased overlap between neighbouring Ti 3d-t<sub>2g</sub> orbitals in the smaller, more distorted cell of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> which accounts

for the closely matched conductivities observed for La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub>.<sup>42, 50</sup> Using the additional data available from the previous report of La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>1-x</sub>Nb<sub>x</sub>O<sub>3</sub>,<sup>8</sup> it is possible to observe the relationship between  $|S|$  and octahedral tilt angle, which is plotted in Fig. 6 d) for compositions with the same nominal carrier concentration and further highlights the effects of tilting on band dispersions discussed above. La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> is cubic ( $Pm\bar{3}m$ ) with zero tilt, while both La<sub>0.55</sub>K<sub>0.45</sub>TiO<sub>3</sub> and La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> are rhombohedral ( $R\bar{3}c$ ) with tilt angles of 2.18(1)° and 7.45(3)° respectively. The magnitude of  $|S|$  increases with increasing tilt angle at all temperatures with a  $\approx 60$   $\mu$ V K<sup>-1</sup> difference between La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> and La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>0.9</sub>Nb<sub>0.1</sub>O<sub>3</sub> due to increased  $m^*$  as a result of reduced orbital overlap brought about by octahedral tilting. In contrast to the clear relationship between  $|S|$  and the tilt angle, similar magnitudes of  $\sigma$  are observed for these materials with the same nominal carrier concentration. From this it is clear that despite targeting high carrier mobility by tuning the symmetry of the unit cell to maximise band dispersions in the zero-tilt structure of La<sub>0.5</sub>K<sub>0.5</sub>TiO<sub>3</sub>, the optimisation of  $\sigma$  is far more complex and depends significantly on the processing conditions of the measured ceramics with extrinsic factors such as grain boundary resistance ultimately dominating  $\sigma$ .<sup>42, 48</sup> The Seebeck coefficient, measured at zero current transport, should be insensitive to grain boundary effects and exhibits a clear correlation to the octahedral tilt angle.

As a result of the improved  $|S|$  values observed for La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> significant differences are observed between the power factors of the A and B site doped materials as shown in Fig. 7 a). Enhancements in  $\sigma$ , particularly at low temperatures, alongside the increased  $m^*$  and  $|S|$  in La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> result in a power factors of 0.58(5) mW m<sup>-1</sup> K<sup>-2</sup> at 600 K compared to 0.39(3) mW m<sup>-1</sup> K<sup>-2</sup> in La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub>. The observation of improved power factors in titanate perovskite thermoelectrics with smaller, more-distorted unit cells is commonplace;  $S^2\sigma$  was enhanced with increasing values of  $x$  in the solid solution Ba<sub>0.9-x</sub>Sr<sub>x</sub>La<sub>0.1</sub>TiO<sub>3</sub>,<sup>51</sup> and the largest  $S^2\sigma$  values were achieved for the smallest lanthanides studied in Sr<sub>0.9</sub>R<sub>0.1</sub>TiO<sub>3- $\delta$</sub> .<sup>42</sup> In this study, when considering both  $\sigma$  and  $S$ , the electronic transport is improved in La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> compositions ( $y = 0.45$  and  $0.4$ ) across the entire measured temperature range compared to B site doped La<sub>0.5</sub>K<sub>0.5</sub>Ti<sub>1-z</sub>Nb<sub>z</sub>O<sub>3</sub> as shown in Fig. 7 a) and is ascribed to the improved Seebeck coefficients which result from lower band dispersions and higher  $m^*$  in the more distorted structures of La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub>. The power factors of the La<sub>1-y</sub>K<sub>y</sub>TiO<sub>3</sub> series are comparable to those of La<sub>0.5</sub>Na<sub>0.5</sub>Ti<sub>1-x</sub>Nb<sub>x</sub>O<sub>3</sub> reported previously.<sup>8</sup>

Even though higher power factors are achieved for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  compared against  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ , both series retain the same glass-like thermal conductivity of the undoped  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  parent and the electronic contribution to the thermal conductivity ( $\kappa_{\text{elec}}$ ) is no larger than 12% for all samples (Fig. S11, ESI). This highlights the weak correlation between electronic and thermal transport within these materials, properties that are usually very strongly coupled in titanate thermoelectrics with  $\kappa_{\text{elec}}$  contributing  $\approx 25\%$  of the total thermal conductivity in similarly doped  $\text{Sr}_{1-x}\text{La}_x\text{TiO}_{3-\delta}$ ,<sup>15</sup> and again demonstrates the advantage of utilising cation mass disorder to control  $\kappa_{\text{latt}}$  which remains the dominant contributor to  $\kappa$  in these materials.

Peak  $ZT$  values of  $\approx 0.18$  are observed at 800 K in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  compositions and the figures of merit increase with temperature for all materials, as shown in Fig. 7 b). These are slightly lower than the highest values obtained previously from  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  and result from the combination of comparable power factors and slightly higher thermal conductivities in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ . The improved performance of A site doped  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  over B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  is clear, not only in peak  $ZT$  but also for the entire measurement range. This increased figure of merit is attributed to the enhanced power factors (Fig. 7a)) which originate from decreased band dispersions in more distorted  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  materials. At all temperatures, the  $ZT$  values of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  are typically 20-50% higher than those of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  with the largest differences in the intermediate temperature range (350-700 K). As a result of this, the average energy conversion efficiencies ( $\varepsilon_{\text{ave}}$ ) of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  are a respectable 1.39-1.59%, for  $y = 0.45$  and  $0.4$  respectively, compared against 0.093-0.097% for in  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  respectively, when evaluated through eqn (9) over the temperature ranges applicable to the recovery of waste heat from automotive exhausts.<sup>4</sup>

$$\varepsilon_{\text{ave}} = \frac{\frac{T_H - T_C}{T_H} \int_{T_C}^{T_H} \frac{\sqrt{ZT+1}-1}{\sqrt{ZT+1} + (T_C/T_H)} dT}{T_H - T_C} \quad (9)$$

The large difference in efficiencies between  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  is due to the increased figures of merit across the temperature range 350-700 K, and not due to the highest peak value. The importance of  $ZT_{\text{average}}$ , particularly for application temperature ranges, is frequently overlooked when searching for new peak  $ZT$  materials.

## Conclusions

This work presents the study of a new solid solution between phonon-glass  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  ( $R\bar{3}c$ ) and  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  ( $Pm\bar{3}m$ ). It is possible to tune the cation size on the A site in order to target the cubic structure with linear Ti–O–Ti bonding that is associated with the high electronic conductivity of  $\text{SrTiO}_3$  based thermoelectrics.  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is the first reported cubic phonon-glass titanate thermoelectric. The phonon-glass thermal conductivity results from localised phonon interactions caused by the disordered distribution of high mass

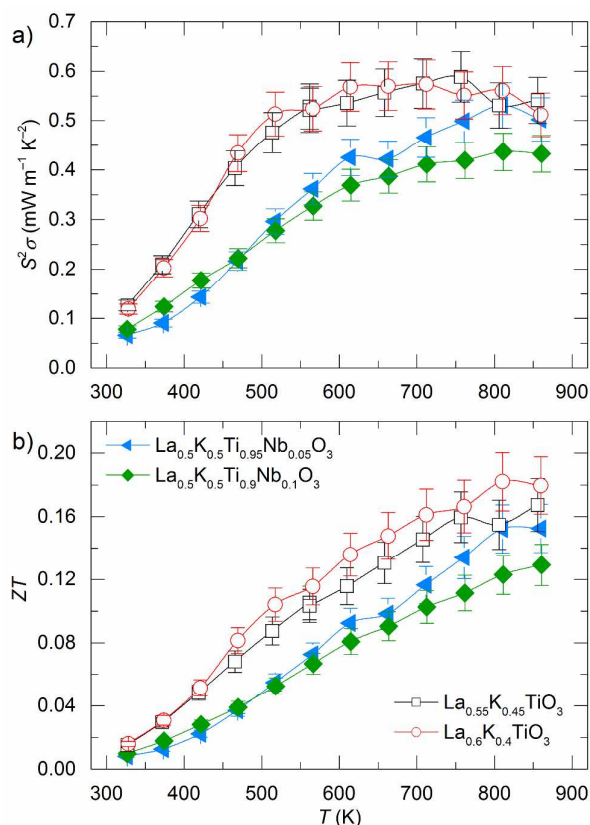


Fig. 7 Temperature dependence of a) thermoelectric power factors ( $S^2\sigma$ ) and b) thermoelectric figures of merit ( $ZT$ ) for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions

contrast  $\text{La}^{3+}$  and  $\text{K}^+$ . Electronic carriers can be introduced through either A or B site doping in  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ , respectively. The A site doping through  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  leads to larger power factors and a higher overall  $ZT$  performance compared to B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ . We find that the electronic conductivity is dominated by extrinsic effects such as grain boundary resistance despite band dispersions being optimised through the cubic symmetry of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The Seebeck coefficient is unaffected by the extrinsic factors and this performance gain is associated with the more distorted and lower symmetry structures of  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  ( $R\bar{3}c$ ) compositions which decrease band dispersion and lead to higher effective masses and Seebeck coefficients compared with  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  ( $Pm\bar{3}m$ ). Phonon-glass thermal conductivities are retained for both A and B site doped materials providing a further example of how intrinsic cation mass disorder can minimise the correlation between the usually strongly coupled thermal and electronic transport in perovskites and yield PGEC behaviour in a cubic titanate thermoelectric. The control of power factors through the distortion of the unit cell could provide guidance in the choice of future target materials, such as combining high mass contrast compositions with smaller cations to target improved electronic transport (e.g.  $\text{Ln}_{0.5}\text{A}_{0.5}\text{TiO}_3$  where  $A = \text{Na}^+/\text{K}^+$ , and  $\text{Ln} = \text{Dy}^{3+}/\text{Ho}^{3+}$ ).



## Conflicts of interest

There are no conflicts to declare.

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**A and B site doping of a phonon-glass perovskite oxide thermoelectric**

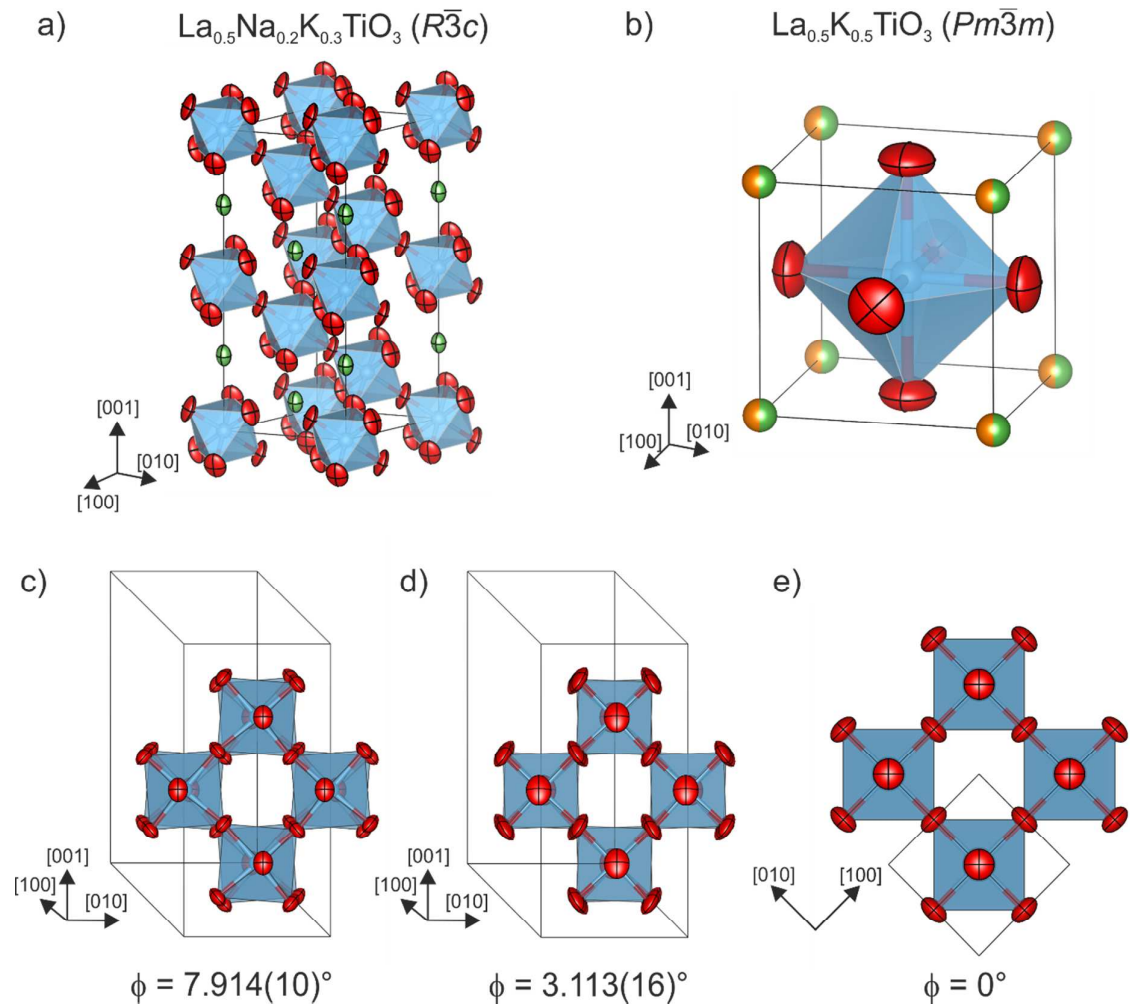
L. M. Daniels, S. Ling, S. N. Savvin, M. J. Pitcher, M. S. Dyer, J. B. Claridge, B. Slater, F. Corà, J. Alaria\* and M. J. Rosseinsky\*

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**Table S1.** Structural parameters of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). O sites are modelled using anisotropic displacement parameters. Space group =  $Pm\bar{3}m$  with dimensions  $a = 3.908212(5) \text{ \AA}$ , and calculated density =  $5.09091(2) \text{ g cm}^{-3}$ . Refinement statistics are  $R_{\text{wp}} = 1.485$ ,  $R_{\text{exp}} = 0.623$ ,  $\chi^2 = 5.667$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	1a	0	0	0	0.4817(8)	0.704(4)
K	1a	0	0	0	0.5183(8)	0.704(4)
Ti	1b	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	1.00	0.644(6)
O	3c	0	$\frac{1}{2}$	$\frac{1}{2}$	0.9994(9)	1.446(16)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
O	0.0066(2)	0.01842(13)	0.01842(13)	0	0	0

\* Symmetry conditions for O site:  $U_{11}, U_{22}=U_{33}, U_{12}=U_{13}=U_{23}=0$



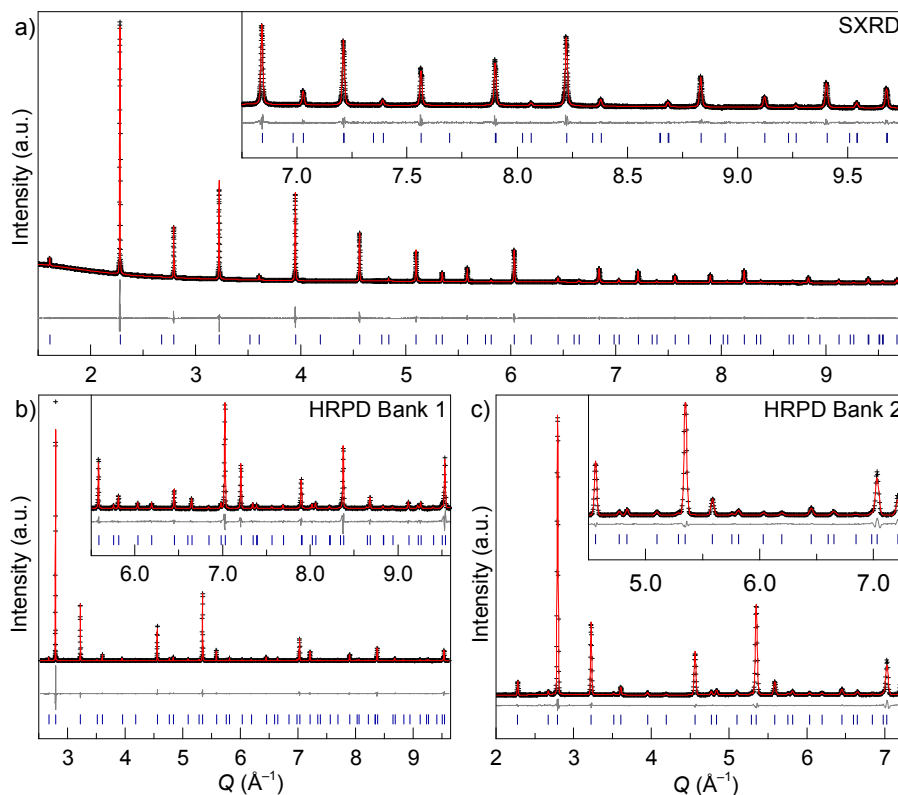
**Fig. S1.** The rhombohedral ( $R\bar{3}c$ ) unit cell of  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  is shown in a), and the cubic ( $Pm\bar{3}m$ ) unit cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is shown in b). A sites are coloured green/yellow, B sites coloured blue and oxygen positions in red. Atom sizes are set by  $U_{\text{iso}}$ . Octahedral tilting in  $x = 0, 0.3$  and  $0.5$  compositions of  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  are shown in c), d) and e), respectively.

**Table S2.** Structural parameters of  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). Coherent neutron scattering lengths: La = 8.24 fm, Na = 3.63 fm, K = 3.67 fm, Ti = -3.438 fm, and O = 5.803 fm.<sup>1</sup> La/Na/K and O sites are modelled using anisotropic displacement parameters. Space group =  $R\bar{3}c$  with dimensions  $a = b = 5.50915(11)$  Å,  $c = 13.4941(6)$  Å,  $\gamma = 120^\circ$ , and calculated density =  $5.0755(2)$  g cm<sup>-3</sup>. Refinement statistics are  $R_{\text{wp}} = 3.004$ ,  $R_{\text{exp}} = 0.623$ ,  $\chi^2 = 23.218$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å <sup>2</sup> ]
La	6a	0	0	$\frac{1}{4}$	0.493(1)	1.15(3)
Na	6a	0	0	$\frac{1}{4}$	0.198(7)	1.15(3)
K	6a	0	0	$\frac{1}{4}$	0.302(7)	1.15(3)
Ti	6b	0	0	0	1.00	0.505(10)
O	18e	0.51564(12)	0	$\frac{1}{4}$	0.999(5)	1.656(16)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
La/Na/K	0.00860(14)	0.00860(14)	0.0172(6)	0.00430(7)	0	0
O	0.0161(8)	0.0082(2)	0.0254(3)	0.0041(8)	-0.0038(4)	-0.0076(9)

\* Symmetry conditions for La/Na/K site:  $U_{11}=U_{22}=2U_{12}$ ,  $U_{33}$ , and  $U_{13}=U_{23}=0$

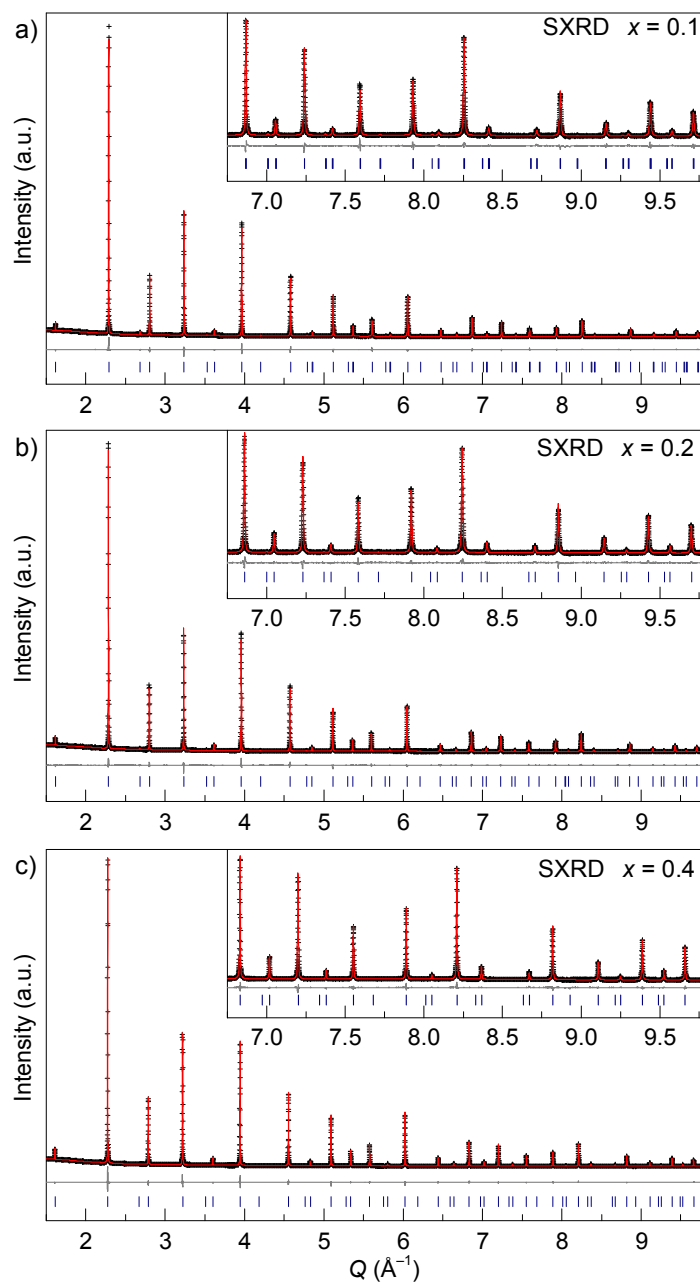
\*\* Symmetry conditions for O site:  $U_{11}$ ,  $U_{22}=2U_{12}$ ,  $U_{33}$ , and  $2U_{13}=U_{23}$



**Fig. S2.** Rietveld refinements against a) SXRD ( $\lambda = 0.82608$  Å) and high resolution neutron powder diffraction (NPD) data measured on b) Bank 1 (backscattering  $2\theta = 168^\circ$ ) and c) Bank 2 ( $2\theta = 90^\circ$  bank) of HRPD at room temperature for  $\text{La}_{0.5}\text{Na}_{0.2}\text{K}_{0.3}\text{TiO}_3$  using space group  $R\bar{3}c$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks.

**Table S3.** Structural parameters of  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  ( $x = 0.1, 0.2$  and  $0.4$ ) series from Rietveld refinement against SXRD data only (298 K).

$x = 0.1, \text{La}_{0.5}\text{Na}_{0.4}\text{K}_{0.1}\text{TiO}_3$						
Space group = $R\bar{3}c$ , $a = b = 5.489273(11) \text{ \AA}$ , $c = 13.43819(6) \text{ \AA}$ , $\gamma = 120^\circ$ , and $\rho_{\text{calc}} = 5.06335(2) \text{ g cm}^{-3}$ , refinement statistics are $R_{\text{wp}} = 2.543$ , $R_{\text{exp}} = 0.941$ , $\chi^2 = 7.311$						
Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	6a	0	0	$\frac{1}{4}$	0.5	0.363(3)
Na	6a	0	0	$\frac{1}{4}$	0.413(3)	0.363(3)
K	6a	0	0	$\frac{1}{4}$	0.087(3)	0.363(3)
Ti	6b	0	0	0	1.0	0.299(4)
O	18e	0.53364(12)	0	$\frac{1}{4}$	1.0	0.759(12)
$x = 0.2, \text{La}_{0.5}\text{Na}_{0.3}\text{K}_{0.2}\text{TiO}_3$						
Space group = $R\bar{3}c$ , $a = b = 5.49564(9) \text{ \AA}$ , $c = 13.4613(5) \text{ \AA}$ , $\gamma = 120^\circ$ , and $\rho_{\text{calc}} = 5.0798(3) \text{ g cm}^{-3}$ , refinement statistics are $R_{\text{wp}} = 2.639$ , $R_{\text{exp}} = 1.109$ , $\chi^2 = 5.675$						
Atom	Wyckoff site	x	Y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	6a	0	0	$\frac{1}{4}$	0.5	0.326(3)
Na	6a	0	0	$\frac{1}{4}$	0.331(3)	0.326(3)
K	6a	0	0	$\frac{1}{4}$	0.169(3)	0.326(3)
Ti	6b	0	0	0	1.0	0.328(4)
O	18e	0.52412(17)	0	$\frac{1}{4}$	1.0	1.103(12)
$x = 0.4, \text{La}_{0.5}\text{Na}_{0.1}\text{K}_{0.4}\text{TiO}_3$						
Space group = $R\bar{3}c$ , $a = b = 5.51816(6) \text{ \AA}$ , $c = 13.5170(3) \text{ \AA}$ , $\gamma = 120^\circ$ , and $\rho_{\text{calc}} = 5.10288(15) \text{ g cm}^{-3}$ , refinement statistics are $R_{\text{wp}} = 2.020$ , $R_{\text{exp}} = 0.850$ , $\chi^2 = 5.641$						
Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}} [\text{\AA}^2]$
La	6a	0	0	$\frac{1}{4}$	0.5	0.658(2)
Na	6a	0	0	$\frac{1}{4}$	0.143(3)	0.658(2)
K	6a	0	0	$\frac{1}{4}$	0.357(3)	0.658(2)
Ti	6b	0	0	0	1.0	0.565(3)
O	18e	0.5069(4)	0	$\frac{1}{4}$	1.0	1.325(12)



**Fig. S3.** Rietveld refinements made against synchrotron X-ray diffraction (SXR D) data ( $\lambda = 0.82608 \text{ \AA}$ ) for a)  $x = 0.1$ , b)  $x = 0.2$  and c)  $x = 0.4$  compositions in the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  solid solution. Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks.

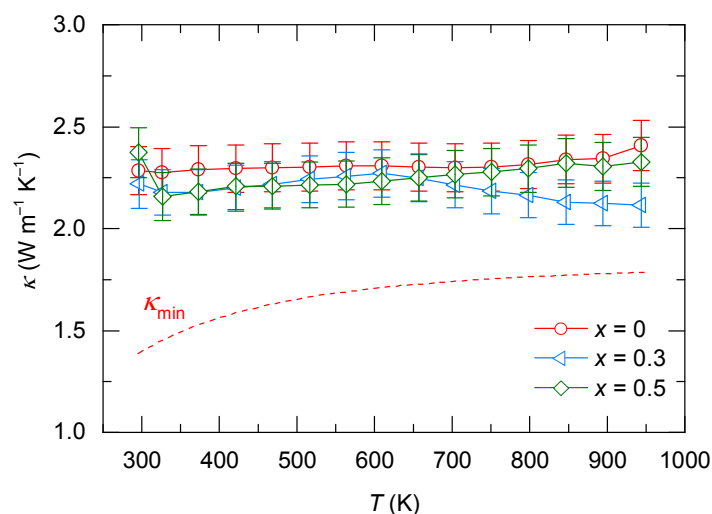
**Table S4.** Bond distances and angles as a function of  $K^+$  substitution in the  $La_{0.5}Na_{0.5-x}K_xTiO_3$  series obtained from Rietveld refinement against room temperature SXR and high resolution NPD data.

$x$	A–O [Å]	Average A–O [Å]	B–O [Å]	A–O/B– O	B–O–B [°]	$\phi$ [°]
0.0	$3 \times 2.5213(3),$ $6 \times 2.74784(3),$ $3 \times 2.9613(3)$	2.74457(17)	1.95003(4)	1.4075	167.044(14)	7.914(10)
0.1	$3 \times 2.5600(6),$ $6 \times 2.74979(6),$ $3 \times 2.9293(6)$	2.7472(3)	1.94915(8)	1.4095	167.13(4)	6.65(2)
0.2	$3 \times 2.6153(9),$ $6 \times 2.75106(6),$ $3 \times 2.8805(9)$	2.7495(5)	1.94757(9)	1.4118	172.19(5)	4.78(3)
0.3	$3 \times 2.6680(4),$ $6 \times 2.75591(2),$ $3 \times 2.8409(4)$	2.7552(2)	1.94965(3)	1.4132	174.92(2)	3.113(16)
0.4	$3 \times 2.721(2),$ $6 \times 2.75939(7),$ $3 \times 2.797(2)$	2.7592(10)	1.95135(10)	1.4140	177.76(13)	1.37(8)
0.5	$12 \times 2.76352$	2.76352	1.95411	1.4142	180	0

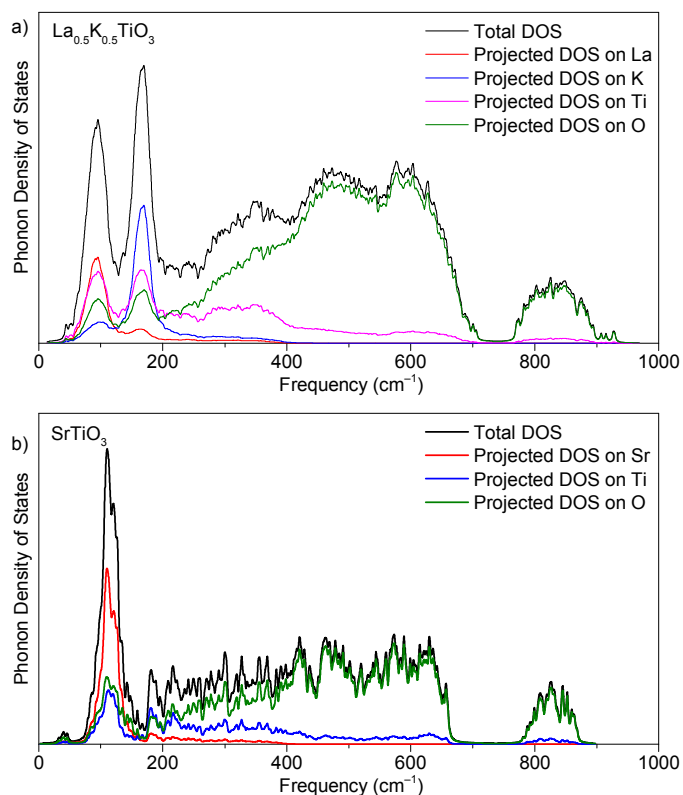
**Table S5.** Parameters used in determination of theoretical minimum thermal conductivity ( $\kappa_{min}$ ) of  $SrTiO_3$  through the Cahill model for disordered crystalline materials.<sup>2</sup> Values of transverse ( $v_{trans}$ ) and longitudinal ( $v_{long}$ ) sound velocities were extracted from single crystals and taken from the literature.<sup>3</sup> These were used to calculate the cut-off frequencies ( $\theta_i$ ) from both through equation (3) in the main text which were then summed as separate contributions to  $\kappa_{min}$  in equation (2).

Material	$v_{trans}$ [ $m s^{-1}$ ]	$\theta_{trans}$ [K]	$v_{long}$ [ $m s^{-1}$ ]	$\theta_{long}$ [K]	$\kappa_{300 K}$ [ $W m^{-1} K^{-1}$ ]	$K_{700 K}$ [ $W m^{-1} K^{-1}$ ]
$SrTiO_3$	4705	612.25	7900	1028.02	1.398	1.741

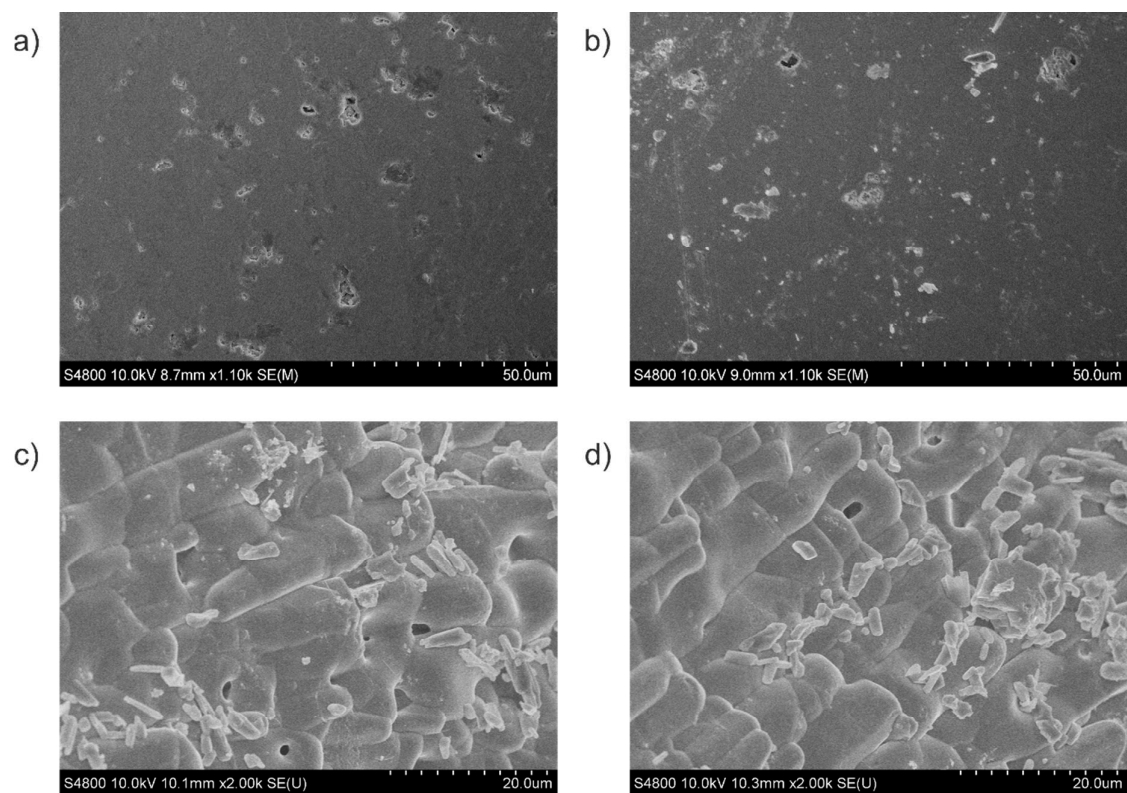




**Fig. S4.** Thermal conductivity ( $\kappa$ ) for  $x = 0, 0.3$  and  $0.5$  in the  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  solid solution highlighting phonon-glass behaviour as a function of temperature. The theoretical minimum thermal conductivity ( $\kappa_{\min}$ ) of  $\text{SrTiO}_3$  is shown as the dashed line and is calculated from Cahill's model for disordered crystalline solids.<sup>2</sup>



**Fig. S5.** Density of states derived from density functional theory (DFT) level phonon calculations for a)  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  and b)  $\text{SrTiO}_3$ , highlighting the excess density of states in the low frequency region for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ .



**Fig. S6.** Scanning electron micrograph images taken from polished and unpolished surfaces of  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  (a & c, respectively) and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  pellets (b & d, respectively).

**Table S6.** Structural parameters of  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  from Rietveld refinement against SXRD data only (298 K). Space group =  $R\bar{3}c$  with dimensions  $a = b = 5.52589(18)$  Å and  $c = 13.5361(8)$ , and calculated density =  $5.28564(19)$  g cm $^{-3}$ . Refinement statistics are  $R_{\text{wp}} = 3.345$ ,  $R_{\text{exp}} = 1.041$ ,  $\chi^2 = 10.375$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å $^2$ ]
La	6a	0	0	$\frac{1}{4}$	0.5494(6)	0.406(3)
K	6a	0	0	$\frac{1}{4}$	0.4506(6)	0.406(3)
Ti	6b	0	0	0	1.0	0.361(4)
O	18e	0.5109(4)	0	$\frac{1}{4}$	1.0	1.483(13)

**Table S7.** Structural parameters of  $\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). La/K and O sites are modelled using anisotropic displacement parameters. Space group =  $R\bar{3}c$  with dimensions  $a = b = 5.52446(2)$  Å and  $c = 13.53672(3)$  Å, and calculated density =  $5.33471(13)$  g cm $^{-3}$ . Refinement statistics are  $R_{\text{wp}} = 3.901$ ,  $R_{\text{exp}} = 1.779$ ,  $\chi^2 = 15.388$ .

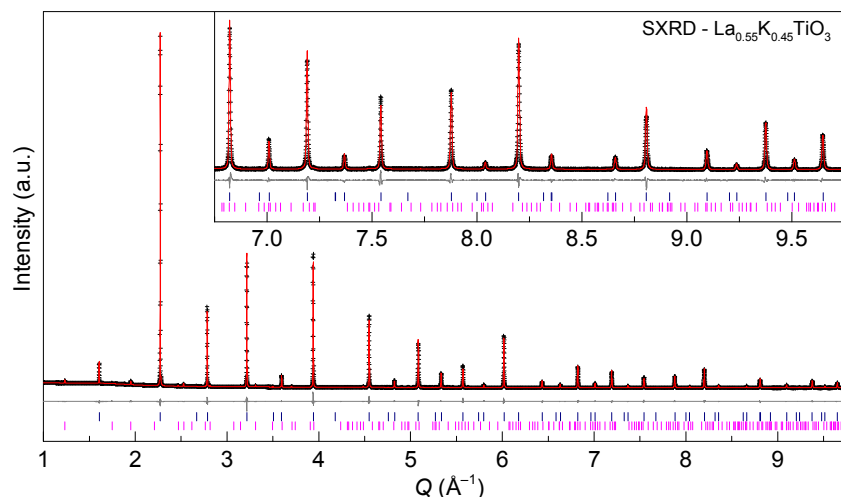
Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å $^2$ ]
La	6a	0	0	$\frac{1}{4}$	0.5786(8)	1.088(9)
K	6a	0	0	$\frac{1}{4}$	0.4214(8)	1.088(9)
Ti	6b	0	0	0	1.00	0.745(16)
O	18e	0.51773(6)	0	$\frac{1}{4}$	1.0005(12)	1.77(9)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
La/K*	0.0045(6)	0.0045(6)	0.0237(16)	0.0022(3)	0	0
O**	0.0159(4)	0.0076(11)	0.0297(12)	0.0038(6)	-0.0048(3)	-0.0097(5)

\* Symmetry conditions for La/K site:  $U_{11}=U_{22}=2U_{12}$ ,  $U_{33}$ , and  $U_{13}=U_{23}=0$

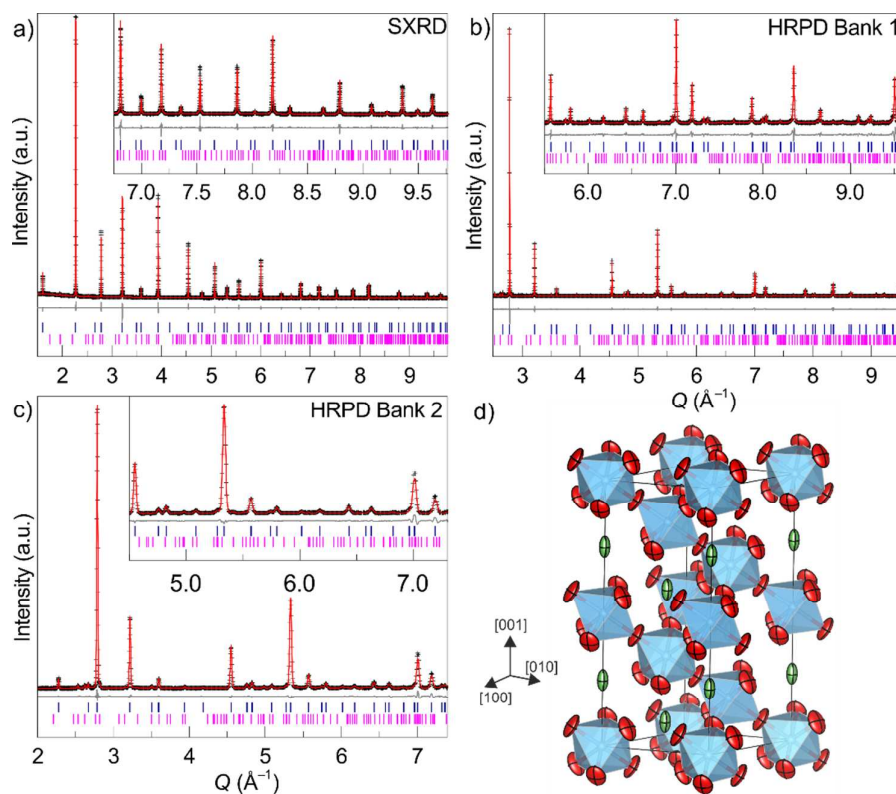
\*\* Symmetry conditions for O site:  $U_{11}$ ,  $U_{22}=2U_{12}$ ,  $U_{33}$ , and  $2U_{13}=U_{23}$

**Table S8.** Bond distances and angles in the  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  series obtained from Rietveld refinement against room temperature SXRD and high resolution NPD data.

y	A–O [Å]	Average A–O [Å]	B–O [Å]	A–O/B–O	B–O–B [°]	$\phi$ [°]
0.45	3 $\times$ 2.7022(17) 6 $\times$ 2.76368(8) 3 $\times$ 2.8237(17)	2.7633(8)	1.95467(11)	1.4137	176.439(10)	2.180(14)
0.4	3 $\times$ 2.6645(3) 6 $\times$ 2.764306(17) 3 $\times$ 2.8604(3)	2.76293(16)	1.95585(2)	1.4127	174.259(19)	3.515(12)



**Fig. S7.** Rietveld refinement against synchrotron X-ray diffraction (SXR) data ( $\lambda = 0.82608$  Å) for  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  using space group  $R\bar{3}c$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up  $< 5\%$  of the sample.



**Fig. S8.** Rietveld refinements against a) SXR ( $\lambda = 0.82608$  Å) and high resolution neutron powder diffraction (NPD) data measured on b) Bank 1 (backscattering  $2\theta = 168^\circ$ ) and c) Bank 2 ( $2\theta = 90^\circ$  bank) of HRPD at room temperature for  $\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$  using space group  $R\bar{3}c$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up  $< 2.5\%$  of the sample. The unit cell of  $\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$  is shown in d) with A sites coloured green, B sites coloured blue and oxygen positions in red.

**Table S9.** Structural parameters of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  from Rietveld refinement against SXRD data only (298 K). Space group =  $Pm\bar{3}m$  with dimensions  $a = 3.912089(3)$  Å, and calculated density =  $5.1850(9)$  g cm<sup>-3</sup>. Refinement statistics are  $R_{\text{wp}} = 3.426$ ,  $R_{\text{exp}} = 1.013$ ,  $\chi^2 = 11.438$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å <sup>2</sup> ]
La	1a	0	0	0	0.5018(7)	0.501(3)
K	1a	0	0	0	0.4982(7)	0.501(3)
Ti	1b	½	½	½	0.9576(8)	0.440(4)
Nb	1b	½	½	½	0.0424(8)	0.440(4)
O	3c	0	½	½	1.0	1.315(12)

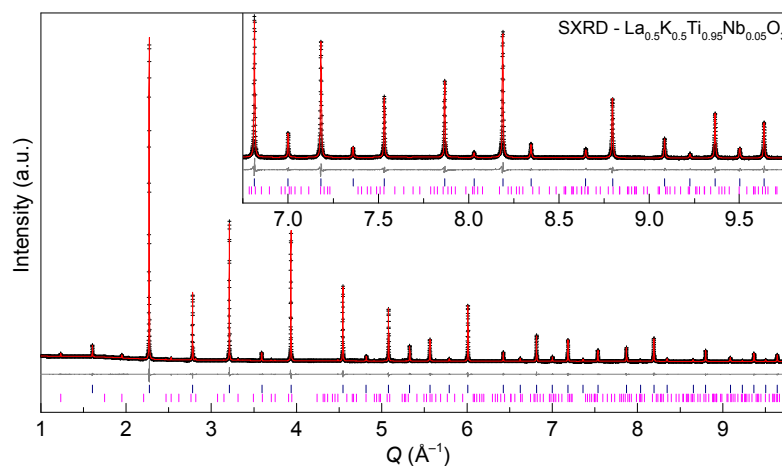
**Table S10.** Structural parameters of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  obtained from Rietveld refinement against SXRD and high resolution NPD data (298 K). Coherent neutron scattering lengths: La = 8.24 fm, K = 3.67 fm, Ti = -3.438 fm, Nb = 7.054 fm and O = 5.803 fm.<sup>†</sup> O sites are modelled using anisotropic displacement parameters. Space group =  $Pm\bar{3}m$  with dimensions  $a = 3.915805(12)$  Å, and calculated density =  $5.22537(5)$  g cm<sup>-3</sup>. Refinement statistics are  $R_{\text{wp}} = 1.809$ ,  $R_{\text{exp}} = 0.719$ ,  $\chi^2 = 6.340$ .

Atom	Wyckoff site	x	y	z	Occupancy	$100 \times U_{\text{iso}}$ [Å <sup>2</sup> ]
La	1a	0	0	0	0.5001(8)	0.777(4)
K	1a	0	0	0	0.4999(8)	0.777(4)
Ti	1b	½	½	½	0.9098(6)	0.682(6)
Nb	1b	½	½	½	0.0902(6)	0.682(6)
O	3c	0	½	½	1.00	1.733(19)
Atom	$U_{11}$	$U_{22}$	$U_{33}$	$U_{12}$	$U_{13}$	$U_{23}$
O	0.0069(3)	0.02252(16)	0.02252(16)	0	0	0

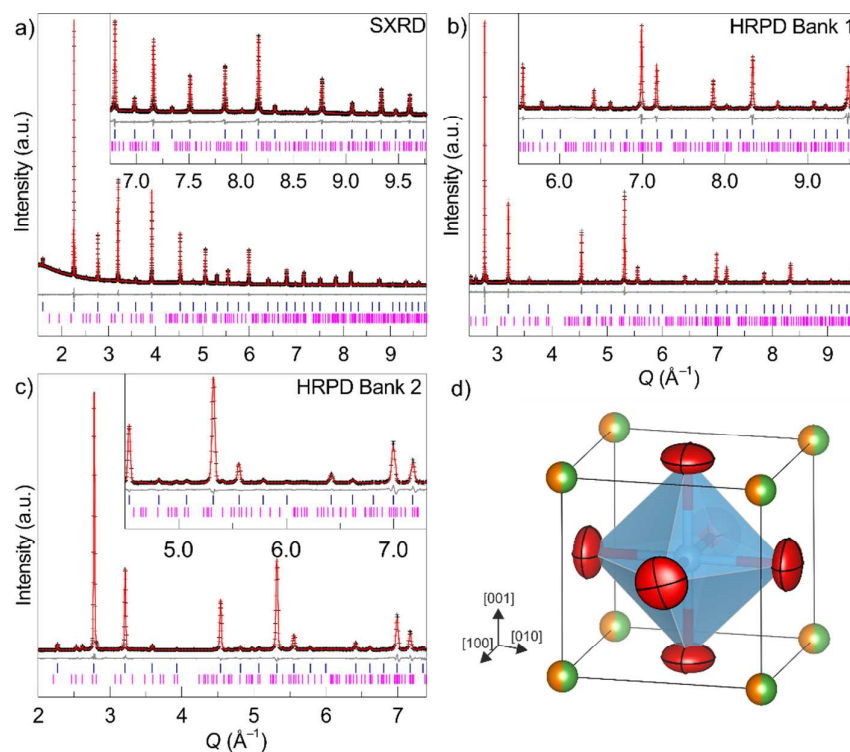
\* Symmetry conditions for O site:  $U_{11}, U_{22}=U_{33}, U_{12}=U_{13}=U_{23}=0$

**Table S11.** Bond distances and angles in the  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  series obtained from Rietveld refinement against room temperature SXRD and high resolution NPD data.

$z$	A–O [Å]	Average A–O [Å]	B–O [Å]	A–O/B–O	B–O–B [°]	$\phi$ [°]
0.05	$12 \times 2.76626$	2.76626	1.95604	1.4142	180	0
0.10	$12 \times 2.76889$	2.76889	1.95790	1.4142	180	0



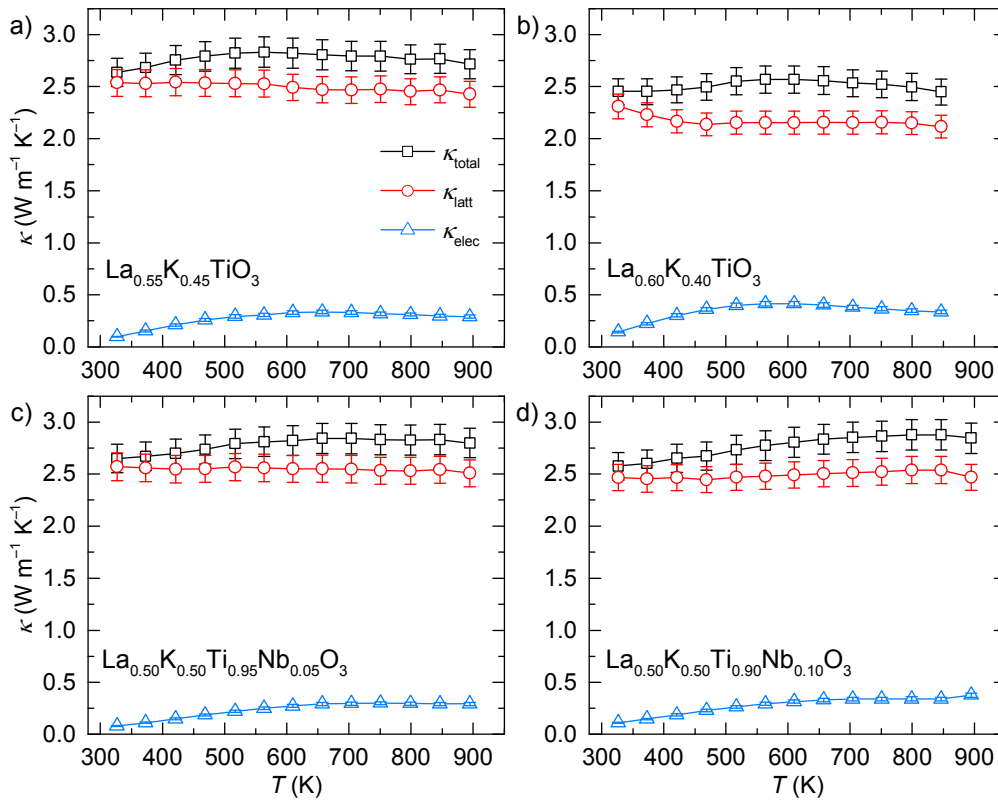
**Fig. S9.** Rietveld refinement against synchrotron X-ray diffraction (SXRD) data ( $\lambda = 0.82608$  Å) for  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$  using space group  $Pm\bar{3}m$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up  $< 5\%$  of the sample.



**Fig. S10.** Rietveld refinements against a) SXRD ( $\lambda = 0.82608$  Å) and high resolution neutron powder diffraction (NPD) data measured on b) Bank 1 (backscattering  $2\theta = 168^\circ$ ) and c) Bank 2 ( $2\theta = 90^\circ$  bank) of HRPD at room temperature for  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  using space group  $Pm\bar{3}m$ . Observed data are shown as black crosses, calculated fit as solid red line, difference as solid grey line, and peak positions as blue tick marks. Pink tick marks show the presence of a  $\text{KTi}_8\text{O}_{16}$  impurity phase which made up  $< 3\%$  of the sample. The unit cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  is shown in d) with A sites coloured green/yellow, B sites coloured blue and oxygen positions in red. Atom sizes are set by  $U_{\text{iso}}$ .

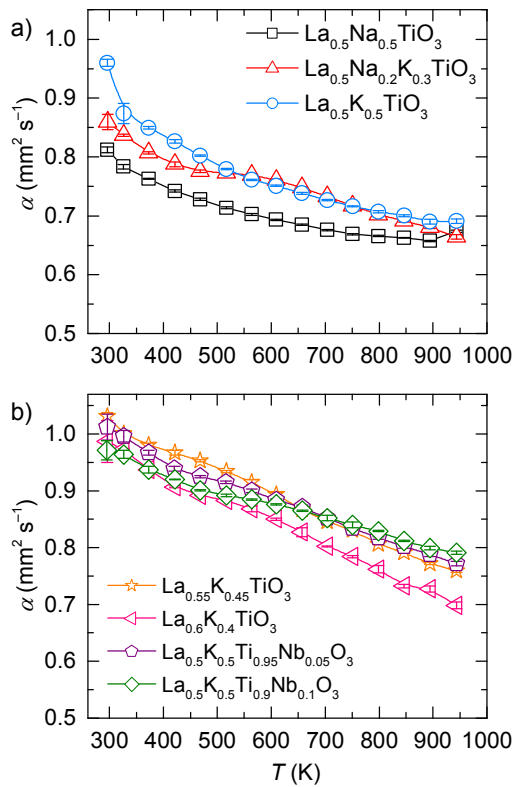
**Table S12.** Theoretical densities ( $\rho_{\text{theo}}$ ) obtained from structural analysis compared with experimental densities ( $\rho_{\text{exp}}$ ) of sintered pellets for each  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  material.

Material	$\rho_{\text{theo}}$ [ $\text{g cm}^{-3}$ ]	$\rho_{\text{exp}}$ [%]
$\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$	5.09091(2)	91.4
$\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$	5.27030(2)	93.9
$\text{La}_{0.6}\text{K}_{0.4}\text{TiO}_3$	5.33471(13)	90.9
$\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.95}\text{Nb}_{0.05}\text{O}_3$	5.1850(9)	96.4
$\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$	5.22537(5)	95.9

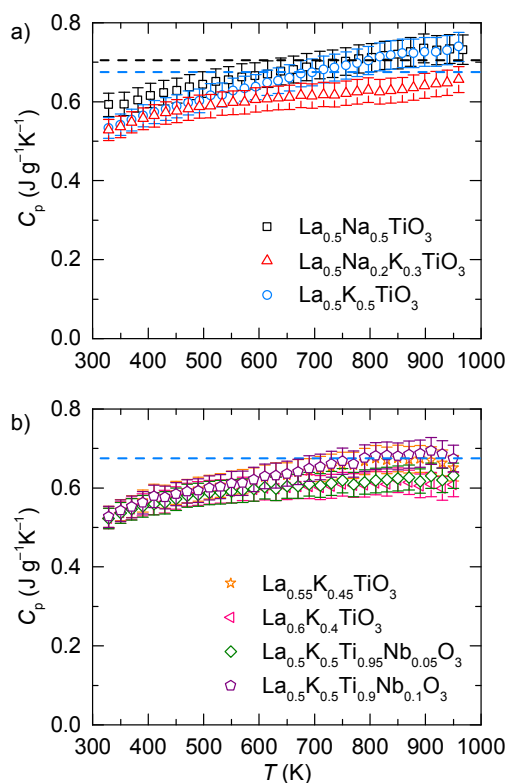


**Fig. S11.** Temperature dependence of  $\kappa_{\text{total}}$  (black squares),  $\kappa_{\text{latt}}$  (red circles), and  $\kappa_{\text{elec}}$  for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  a)  $y = 0.45$ , b)  $y = 0.4$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  c)  $z = 0.05$ , and d)  $z = 0.1$ . The electronic contributions to the thermal conductivity were calculated through the Wiedemann-Franz law.<sup>4</sup>





**Fig. S12.** Temperature dependence of the thermal diffusivity ( $\alpha$ ) for a)  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  and b) reduced  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions.



**Fig. S13.** Temperature dependence of the specific heat capacity ( $C_p$ ) for a)  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  and b) reduced  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_2\text{O}_3$  compositions. Error bars show 5% uncertainty. Dashed lines represent  $C_p$  values for  $\text{La}_{0.5}\text{Na}_{0.5}\text{TiO}_3$  (black) and  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  (blue) estimated from the Dulong-Petit law,  $C_p = 3nR$ , where  $R$  is the molar gas constant,  $8.314 \text{ J K}^{-1} \text{ mol}^{-1}$  and  $n$  is the number of atoms per unit volume.

## References

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Nazario Martin

Editor in Chief

Journal of Materials Chemistry A

Dear Professor Martin,

**A and B Site Doping of Phonon-glass Perovskite Oxide Thermoelectric**

Please consider this article for publication in Journal of Materials Chemistry A.

The paper reports the use of simple substitutional and crystal chemistry in order to target the first phonon-glass electron-crystal (PGEC) thermoelectric perovskite oxide with cubic symmetry. This is significant due to the current interest in high temperature energy recovery in the automotive and industrial sectors using thermally robust and non-toxic oxide materials. The PGEC concept envisages the simultaneous optimisation of thermal conductivity, electronic conductivity and thermopower to maximise the thermoelectric figure of merit (ZT), and is achieved by combining the thermal properties of a glass with the electronic transport properties of a crystalline material. We recently reported the first PGEC oxides (*Energy and Environmental Science*, 10 (2017) 1917) and it is important to understand how the properties of this new class of materials can be chemically controlled. Here we investigate the effect of the symmetry of the materials on the thermoelectric properties.

The study uses theoretical calculations, which we developed recently in order to fully understand the physical mechanisms responsible for PGEC behaviour in  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$ , as predictors to identify desirable structural features that lead to such properties. Using the flexibility of the perovskite structure and simple cation substitution, we show how it is possible to tune the electronic structure such that band dispersions are optimised towards high electronic conductivity as predicted computationally. Phonon calculations showed that the disordered mixtures of A site metals in these compounds yielded localised phonons similar to those observed previously in  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$ . Under guidance of theoretical calculations, a perovskite oxide with cubic symmetry optimised towards high carrier mobility and a phonon-glass thermal conductivity is realised experimentally. This shows how our computational approach can be used to direct the synthesis of new materials optimised towards high thermoelectric performance.

By maintaining the general  $\text{ABO}_3$  perovskite stoichiometry, it is possible to dope electronic carriers through either A or B site substitution and retain the phonon-glass thermal conductivity to achieve decoupling of thermal and electronic transport in these materials. We show that by using precise control of octahedral distortion which has a clear correlation to electronic transport, the A site doping is ultimately more beneficial towards obtaining the highest thermoelectric power factor in these phonon-glass thermoelectric oxides, with substantial performance gains over B site doping. This control of electronic transport through precise structural distortion and control of thermal transport

through the introduction of cation and mass disorder presents a significant step forward in oxide thermoelectrics and is a technique that can be applied to a wide range of materials, and can be considered in the search for future target oxide thermoelectric materials.

Please do not hesitate to contact me if you require any further information.

Yours sincerely

Matthew Rosseinsky

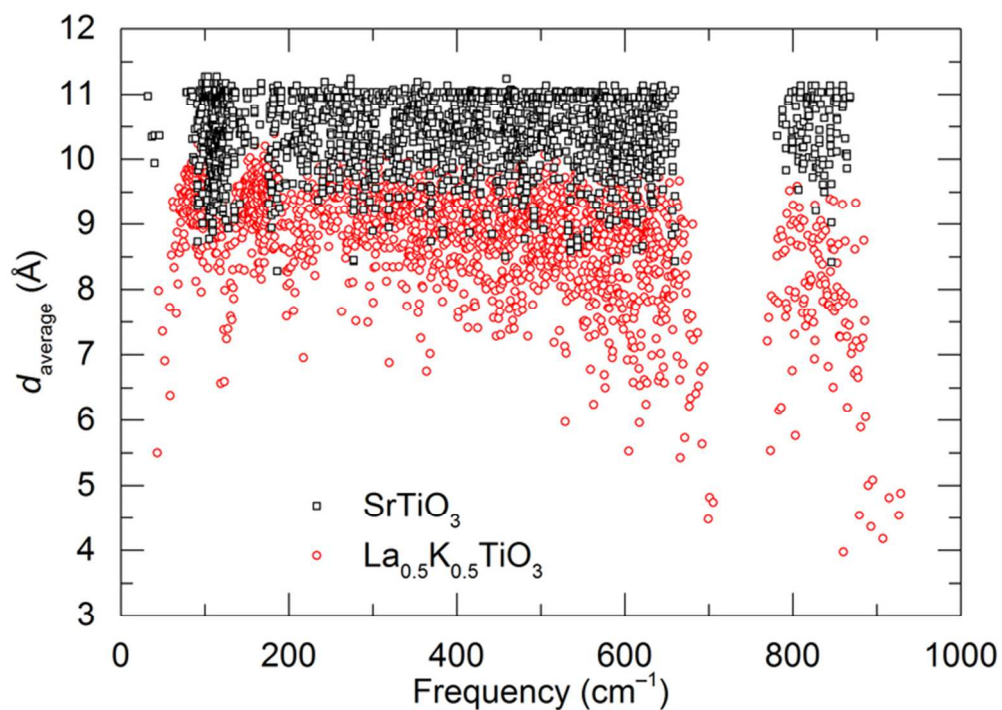


Fig. 1 Phonon localisation in  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  (red circles) determined from DFT calculations compared against interactions in the phonon-crystal  $\text{SrTiO}_3$  (black squares). The parameter  $d_{\text{average}}$  is calculated for individual phonon modes as the mean distance between the ten atoms that participate to the greatest extent in that phonon mode as defined by their displacement amplitudes.

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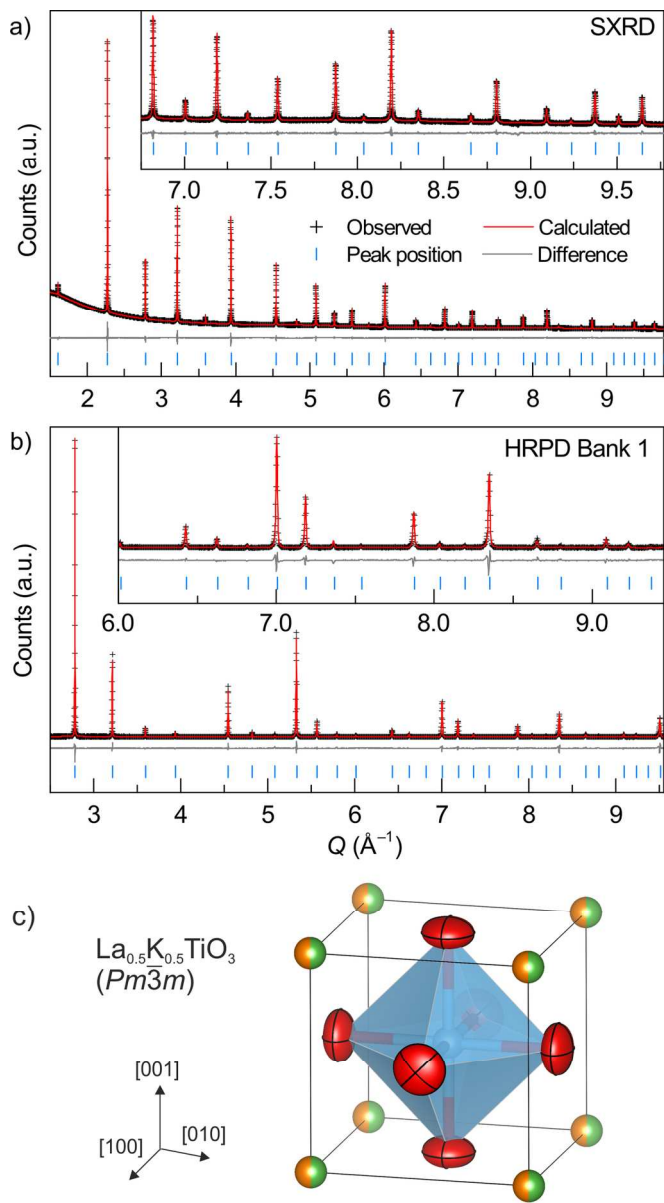


Fig. 2 Rietveld refinements against a) SXR D ( $\lambda = 0.82608 \text{ \AA}$ ) and b) high resolution NPD data measured on Bank 1 (backscattering  $2\theta = 168^\circ$ ) of HRPD at room temperature for  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . The refinement model is based on cubic  $Pm\bar{3}m$  symmetry with A site metals  $\text{La}^{3+}$  and  $\text{K}^+$  modelled on the 1a (0, 0, 0) position, the  $\text{Ti}^{4+}$  on the 1b ( $\frac{1}{2}$ ,  $\frac{1}{2}$ ,  $\frac{1}{2}$ ) position and oxide ions on the 3c positions (0,  $\frac{1}{2}$ ,  $\frac{1}{2}$ ). The cubic cell of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$  is shown in c) with anisotropic displacement ellipsoids for the oxide positions.

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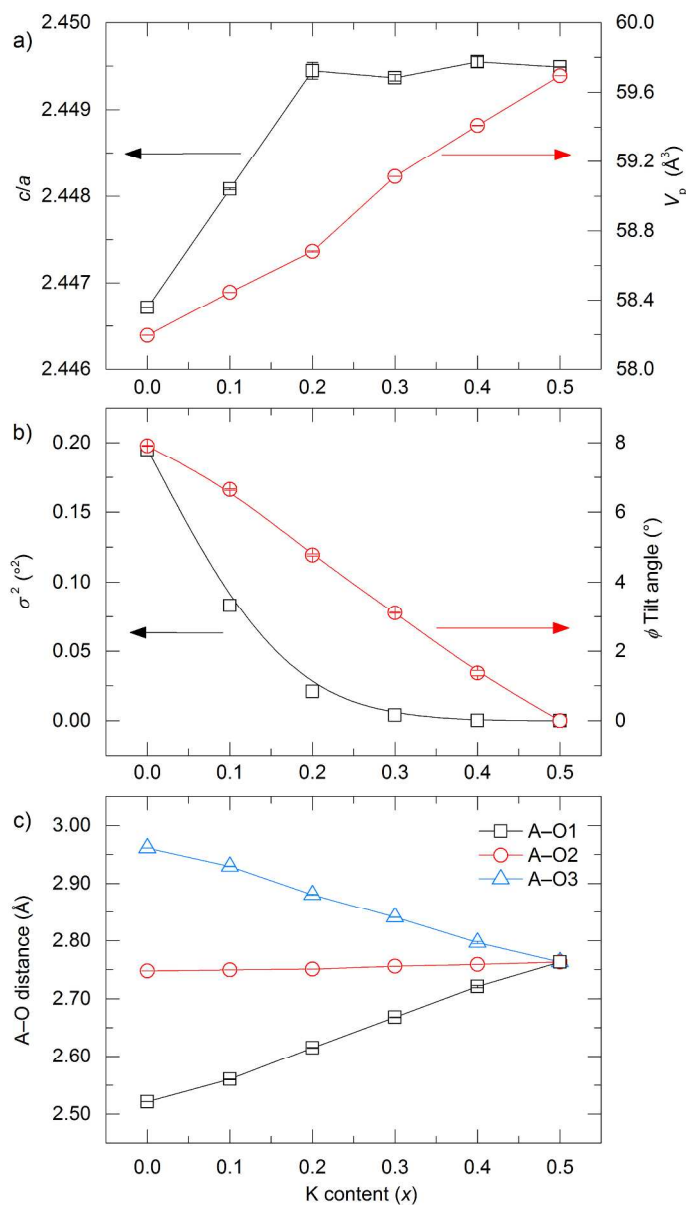


Fig. 3 Variation in a) c/a lattice parameter ratio (black squares) and primitive pseudocubic cell volume (red circles), b) bond angle variance (black squares) and octahedral tilt angle (red circles), and c) A–O distances as a function of x in the solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$  refined from SXRD and high resolution NPD data.

Primitive cell parameters are obtained through  $a_p = a/\sqrt{2}$  and  $c_p = c/2\sqrt{3}$ . Octahedral tilt angles ( $\phi$ ) are determined through  $\phi = \arctan[2\sqrt{3}(x-0.5)]$ , where x is the x coordinate of the 18e oxide position in space group  $R\bar{3}C$ . Bond angle variance is obtained from eqn (2) in the text.

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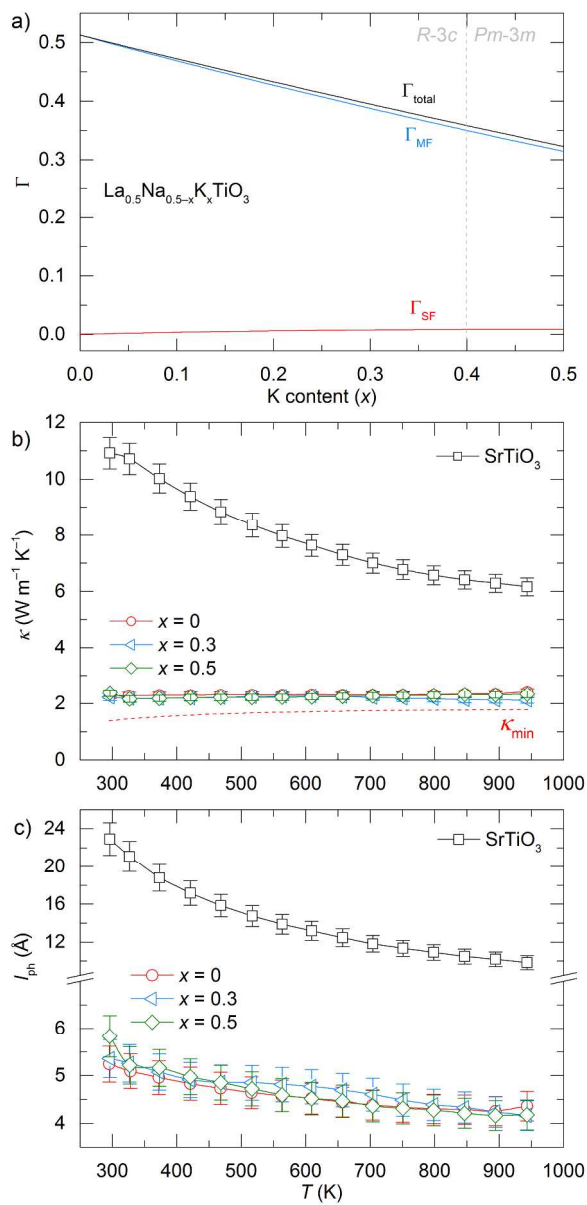


Fig. 4 Dependence of a) disorder scattering parameter ( $\Gamma$ ), including mass ( $\Gamma_{\text{MF}}$ ) and size ( $\Gamma_{\text{SF}}$ ) contributions shown in blue and red respectively, b) thermal conductivity ( $\kappa$ ), and c) phonon mean free path ( $l_{\text{ph}}$ ) on  $x$  in the solid solution  $\text{La}_{0.5}\text{Na}_{0.5-x}\text{K}_x\text{TiO}_3$ . The theoretical minimum thermal conductivity ( $\kappa_{\text{min}}$ ) of  $\text{SrTiO}_3$  is shown as the dashed line in b) and is calculated from Cahill's model for disordered crystalline solids.[REF-Cahill1992] Phonon mean free paths in c) are obtained through eqns (4) and (5).

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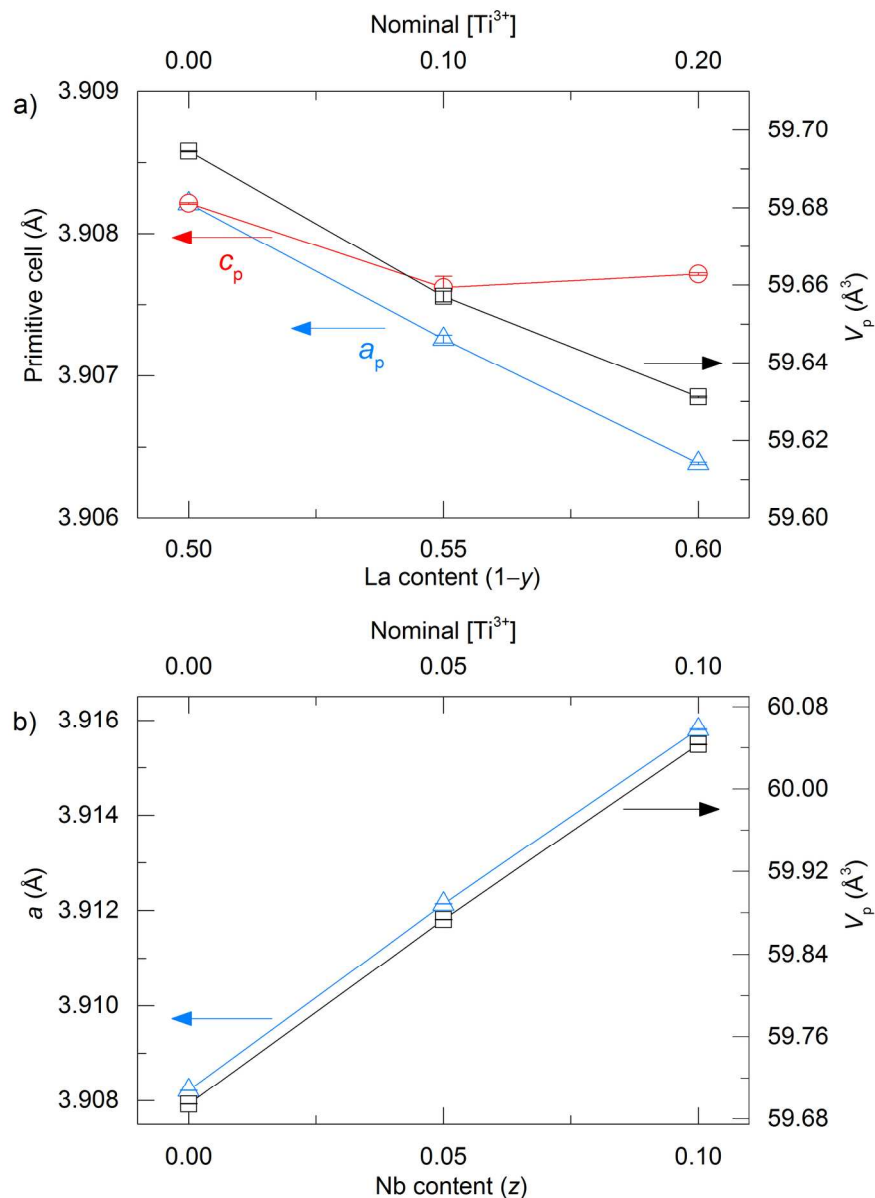


Fig. 5 Variation in a) primitive unit cell parameters and volume as a function of  $\text{La}^{3+}$  content in A site doped  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  with  $R\bar{3}c$  symmetry, and b) cubic unit cell parameter and volume in B site doped  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  ( $\text{Pm}\bar{3}m$ ) as a function  $\text{Nb}^{5+}$  substitution. The nominal  $\text{Ti}^{3+}$  concentrations are plotted on the top axes of each and increase with increasing  $\text{La}^{3+}$  and  $\text{Nb}^{5+}$  content. The  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  series are cubic ( $\text{Pm}\bar{3}m$ ), while  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  materials are rhombohedral ( $R\bar{3}c$ ).

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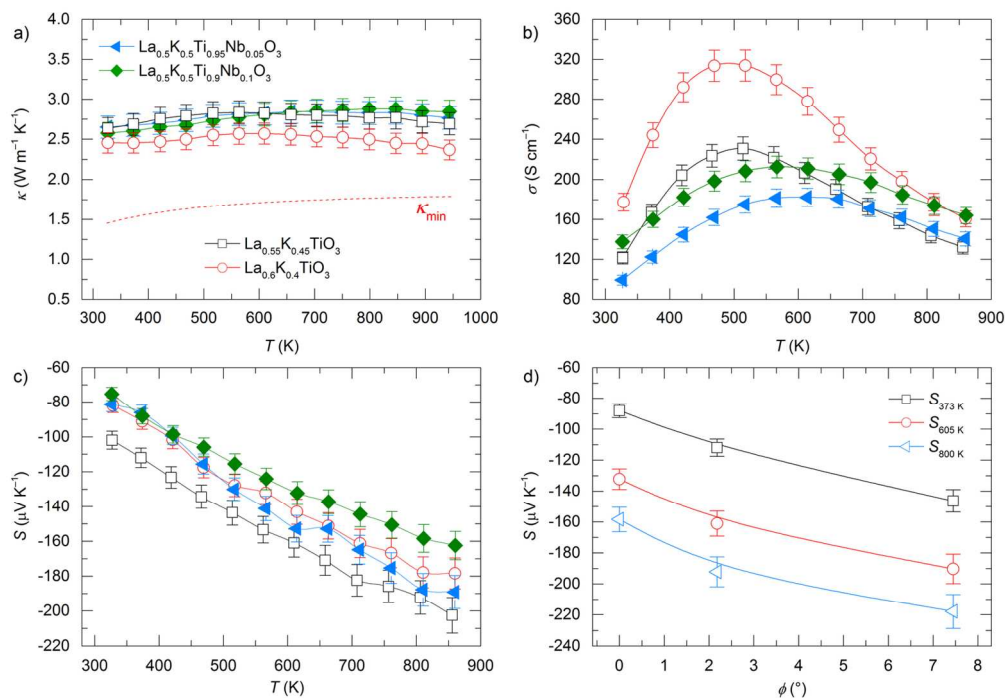


Fig. 6 Temperature dependence of a) thermal conductivity ( $\kappa$ ), b) electronic conductivity ( $\sigma$ ), and c) Seebeck coefficient ( $S$ ) for A site ( $\text{La}_{1-y}\text{K}_y\text{TiO}_3$ ,  $y = 0.45$  and  $0.4$ ) and B site ( $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$ ,  $z = 0.05$  and  $0.1$ ) doped compositions of  $\text{La}_{0.5}\text{K}_{0.5}\text{TiO}_3$ . Dependence of  $S$  as a function of tilt angle ( $\phi$ ) at different temperatures is shown in d) using three materials that have the same nominal carrier concentration:  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  ( $\phi = 0^\circ$ ) and  $\text{La}_{0.55}\text{K}_{0.45}\text{TiO}_3$  ( $\phi = 2.18(1)^\circ$ ) from this study, and  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{0.9}\text{Nb}_{0.1}\text{O}_3$  ( $\phi = 7.45(3)^\circ$ ) which was taken from a recent study of PGE  $\text{La}_{0.5}\text{Na}_{0.5}\text{Ti}_{1-x}\text{Nb}_x\text{O}_3$  materials with  $R\bar{3}c$  symmetry. [REF-Daniels2017]

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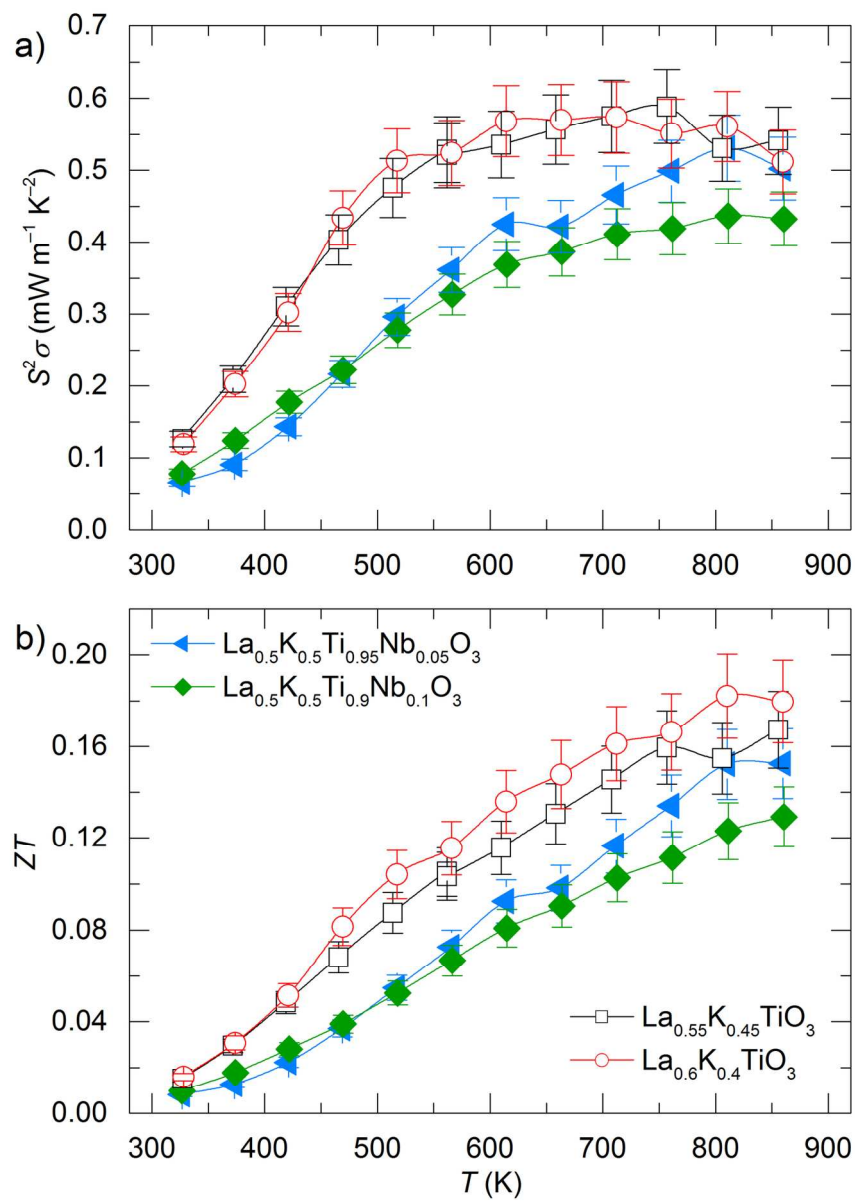


Fig. 7 Temperature dependence of a) thermoelectric power factors ( $S^2\sigma$ ) and b) thermoelectric figures of merit ( $ZT$ ) for  $\text{La}_{1-y}\text{K}_y\text{TiO}_3$  and  $\text{La}_{0.5}\text{K}_{0.5}\text{Ti}_{1-z}\text{Nb}_z\text{O}_3$  compositions

133x186mm (300 x 300 DPI)